METALLURGIA

The British Journal of Metals

(INCORPORATING THE METALLURGICAL ENGINEER)

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METALLURGIA

THE BRITISH JOURNAL OF METALS.

INCORPORATING "THE METALLURGICAL ENGINEER."

MAY, 1943 VOL. XXVIII. No. 163

Fast Cargo Ships?

ALTHOUGH in recent months shipping losses have been greatly reduced, the war at sea continues, and the fight goes on without pause. Naturally, considerable attention has been given to the problem of combating this menace to our supply lines, and the improvement in the position can be regarded as due to the success of the measures applied. Undoubtedly the increased use of aircraft has and is being effective in reducing the activities of U-boats, though other measures applied by surface vessels have evidently contributed in no small degree.

On several occasions, especially during the most difficult periods, criticism of the measures applied to combat the U-boats appeared in the press and the need for faster cargo ships was stressed many times. Indeed, only a week or two ago, it was stated by Admiral Sir Hugh Tweedie that with a fast, well-found convoy there would be no stragglers. The convoy can zig-zag at speed, constituting a real danger to the submarine itself, for to get near the convoy the submarine must operate on the surface, exposing herself to attack from the air or surface escort. Moreover, by steaming faster, vessels are fewer days at sea, giving greater concentration of escort without having to divert slow-laden ships half round the world in order that they may enjoy what protection is possible as they approach these islands.

It is not proposed to enter into any argument on the subject, though a fast convoy would also have its stragglers, but it is probably true to say that if all the ships used by the Allied countries for supplies had a minimum speed of 15 knots the position would have been greatly simplified. Unfortunately, however, at the outbreak of war the average speed of cargo vessels would not exceed 11 knots, obviously therefore the convoy system had to be built up with vessels that were available and these had to be supplemented by new ships as fast as the shipyards could construct them. To have embarked upon the construction of, say, 15-knot ships, to replace losses, would have involved considerable reorganisation of the shipyards and marine engineering works at a time when delay would have been suicidal.

In normal times the economical speed of a cargo ship depends upon a rather sensitive balance between several factors, such as earnings from cargo carried, loading and discharging times and cost, running and capital charges, insurance, etc. Before the war progress in marine engineering and in shipbuilding was altering this balance; the cost of the development of power, for instance, was steadily diminishing, in relation to other charges, and the speed was gradually being increased to maintain the balance required. New methods of hull construction and general improvement in hull form have improved the deadweight ratio and the average speed. Changes were being made slowly to conform with economical considerations, and although under war conditions other considerations are involved, the adoption of revolutionary changes absorbs a tremendous amount of time before construction can proceed at the speed desired.

It must be remembered that what is desired in war time is the transport of the maximum tonnage in a given time. In this connection, it is interesting to note that Mr. J. Ramsay Gebbie, O.B.E., president of the North-East Coast Institution of Engineers and Shipbuilders, recently examined the possibilities of transporting the largest amount of cargo, the conditions being associated with the output of a marine-engine firm which had a yearly production of twenty-four 2,900 i.h.p. oil engines suitable for 11-knot cargo ships, or, alternatively, of ten 7,000 i.h.p. oil engines suitable for 15-knot ships of the same dimensions; the object being to ascertain whether it is the fast or the less fast ship which accomplishes this.

He assumed that a start is made from scratch and that the output of finished ships is entirely dependent on the supply of machinery and his comparisons are made for motor-ships. The ships compared are: (1) Motor-ships of 10,300 tons deadweight with engines of 2,900 i.h.p. and having a speed of 11¼ knots voyage average on a consumption of 10 tons of oil-fuel per 24 hours; and (2) Motor-ships of the same dimensions which with their finer model and heavier machinery would carry 8,300 tons deadweight when fitted with engines of 7,500 i.h.p. giving them a speed of 15 knots on a consumption of 27 tons per 24 hours.

It has been assumed that either type would be six months on the stocks, but that the 11¼-knot ship would take two months to fit out and the 15-knot ship three months.

From figures given, it is shown that at the end of three years from the commencement of the alternative programmes fifty-eight 11¼-knot ships of 10,300 tons deadweight would have been delivered as against twenty-three 15-knot ships of 8,300 tons deadweight, and that the slower vessels would have carried 6,162,200 tons of cargo as against 2,142,800 tons carried by the faster ships, assuming round voyages of 10,000 miles; in other words, it would appear that the fifty-eight 11¼-knot vessels would transport nearly three times the weight of cargo carried by the twenty-three 15-knot ships.

In the comparison, it has been assumed that the materials for making the 7,500 i.h.p. machinery would be available and forthcoming at the proper times, but it should be remembered that when the horse-power of an engine is increased about 2½ times the size and weight of all forgings and castings, plates, gearing,

The fact that goods made of raw materials in short supply owing to war conditions are advertised in "Metallurgia" should not be taken as an indication that they are necessarily available for export.

and indeed all construction parts are considerably increased.

It is probable that in a few years' time the 15-knot cargo ship will be in general use, but the problem of making a revolutionary change in speed in war time is not so simple as it appears to be at first sight. It will be appreciated that there can be no hold-up in the transport of supplies, and to achieve this the increased number of the slower vessels which can be built in a given time permits the transport of more cargo in that time.

Industrial Bottlenecks

OPINIONS will always differ on what constitutes the most serious bottlenecks to the maintenance of a high output for the successful prosecution of the war; but Senator Truman, chairman of the Senate Committee investigating bottlenecks in American war industries, has no doubts on this subject. As a result of an independent study of the problem he concludes that the human factor is the chief difficulty, that suspicions, rivalry, apathy and greed lie behind most bottlenecks. After giving much thought to this matter, he believes the fact must be faced that while many so-called industrial incentive schemes are good for occasional spurts, they are apt to fade in effectiveness with time. This can only be overcome by instilling the personal qualities of patriotism and self-sacrifice based on moral and spiritual principles.

In this country it is probable that peak production per worker hour was reached following the fall of France, and the withdrawal of our forces from Dunkirk, and it is to be hoped that the result of the great victory in North Africa will have the same effect.

This Issue

DESPITE continued efforts to make the best possible use of our quota of paper, while still retaining the normal size of page, the gradual but persistent increase in the demand for METALLURGIA has forced us to make a reduction in its size. We have been reluctant to make this change, but by reducing to the size of the present issue a reasonable saving can be effected which will permit us to further improve our service.

We trust that the change will be agreeable to all who use the journal, and that, in its new form, it may be found convenient to handle and to maintain the high standard associated with it in the past.

"The Casting of Brass Ingots"

PUBLISHED in 1934 by the British Non-Ferrous Metals Research Association, this book enjoyed a wide circulation. For obvious reasons, it is now in considerable demand, and as the original is out of print, the Association is reissuing it by the photo-litho-offset process, complete with all illustrations as in the original edition. The book was based on a comprehensive series of investigations carried out by the authors for the Association, and as the fundamental principles emerging from these investigations remain unchallenged, no revision has been undertaken: it is felt that in its original form the book will be a valuable aid to industry, particularly to individuals and companies, unaccustomed to the technique of brass casting, who have turned under war conditions to this branch of metallurgical practice. The new issue (191 pp., 123 figures) will appear shortly, price 21s. Orders may be placed with the B.N.F.M.R.A., Euston Street, London, N.W. 1, or with any bookseller.

The Imperial Institute's 50th Anniversary

THE Imperial Institute was founded in 1887 as a memorial of Queen Victoria's golden jubilee, but the building housing the Institute was officially opened in May, 1893, thus the present month celebrates its 50th anniversary of service. After many vicissitudes, n 1903 the management was vested in the Board of Trade, assisted by an Advisory Committee, which included representatives of India and the Colonies; in 1925 the Imperial Institute Act was passed giving the Institute its present constitution. By the terms of this Act the Secretary of the Department of Overseas Trade was nominated as the Minister responsible to Parliament for the affairs of the Institute. He exercises his responsibilities with the aid of a Board of Governors, of whom he is the president, and who comprise representatives of the Dominions of India, the Treasury, the Board of Trade, the Colonial Office, the Ministry of Agriculture and Fisheries, and the Department of Scientific and Industrial Research, as well as representatives of scientific and commercial interests.

The general purposes of the Institute are defined in the 1925 Act of Parliament, which directs that the Institute is to be used to promote the commercial, industrial and educational interests of the British Empire; to collect and disseminate information upon a variety of subjects relating to the marketing and uses of the raw materials or semi-manufactured products of the Empire and in general to supply technical and scientific information bearing upon the industries of the Empire; to advise on the development of the resources of the Empire in raw materials in order that such resources may be made available for the purposes of industry and commerce and of Imperial defence; and to conduct in the laboratories of the Institute preliminary investigations of raw materials.

The services of the Imperial Institute are widespread throughout the Empire; they extend to every Government of the Empire, Home and Overseas, as well as to producers, merchants and manufacturers. They include not only the science of technical information to promote the development of the resources of the Empire in raw materials, but also the art of visual instruction in the life, scenery and industries of the Overseas Empire. Experience has shown, however, that many are not aware of the extensive facilities available at the Institute for the rapid supply of technical information relating to the trade, occurrence and utilisation throughout the world of all kinds of raw materials. The scope of the intelligence service is also not so well known to commercial firms as it might be. It should be remembered that the Institute's staff includes tropical agriculturists, chemists, chemical technologists, economic botanists, economic geologists, mining engineers, mineralogists and statisticians, all of whom are expert in their particular subjects.

The facilities of this Institute should be drawn upon to an ever-increasing extent, especially in this time of emergency; arrangements have been in force for some time past for dealing with inquiries as speedily as possible and, wherever expedient, by telephone. Indeed, there is no doubt that the Director, Sir Harry Lindsay, K.C.I.E., C.B.E., would welcome increased inquiries as an appreciation of the useful work performed by the Institute in its 50 years' service to the Empire.

The Hardness Testing of Metals and Alloys

By J. W. Donaldson, D.Sc.

The property of hardness is generally appreciated though not easily measured. Much work has been carried out and much has been written on this property in its connection with engineering materials, and to-day many methods are applied to measure the hardness of metals and alloys. Opinions differ on what constitutes the most satisfactory means. In this review the author discusses some factors involved in the testing of this property and directs particular attention to methods applied to the testing of metals and alloys.

THE most frequently applied test for the control of the quality of engineering materials and metallurgical products is the hardness test. The principal advantage of this method of testing is the facility with which it can be easily and rapidly carried out, in that it does not involve much preparation of the material for testing, and the time required for testing is comparatively short. The hardness test is usually a non-destructive test and does not involve waste of material for test purposes. The hardness test also not only affords a considerable amount of information regarding the strength properties of material, particularly with regard to steel where most hardness measurements can be correlated with tensile strength, but it is also a means of checking the uniformity of material and the accuracy and success of heat-treatment operations. In this latter respect it is a most useful test to engineers and metallurgists.

Methods of testing and measuring the hardness of metals and alloys in modern use can be divided into scratch tests, static tests and dynamic tests. The best known scratch test is that devised by Mohs for minerals, but methods using a diamond point on a polished metallic surface have also been developed by Turner and Martens. Hardness is measured in the former of these two tests by the load required to produce a visible scratch, and in the latter by the load required to produce a scratch 0·01 mm. in breadth. Scratch tests are not used to any great extent in modern engineering and metallurgical practice. Static tests are the most important hardness tests, and include all indentation methods, such as are used in the Brinell, Pyramid Hardness, Rockwell and Monotron testing machines. Indentations in those tests are produced by a sphere, prism, cone or cylinder.

Dynamic methods of hardness testing include rebound test methods—methods employing penetration of a steel ball forced into the test specimen by impact with subsequent measurement of the penetration—and the pendulum process of testing devised by Herbert. The Shore scleroscope is the best known example of the rebound hardness test, but another instrument of this type is the Duroscope, which works on the pendulum principle. The Herbert compound pendulum is different in principle and action from any other hardness testing instrument, and can be used to measure other properties of metals besides hardness.

In the wide variety of instruments and machines developed for hardness testing, not only have the more generally used of those instruments been standardised, but many improvements and modifications have been made during recent years. Some of these modifications are not in common use and have been developed in order to meet difficulties encountered by users of hardness testing equipment, such as the testing of round and cylindrical specimens, the testing of thin materials and the testing of micro-constituents of metals and alloys. Attempts have also been made to form conversion tables so that different methods of hardness testing can be directly comparable. A review, therefore, of the various hardness tests in general use, the modifications which have been developed to meet special conditions, and a consideration of the various hardness comparison scales which have been established is of interest and use to all users of metals and alloys.

Brinell Hardness Tests

The Brinell hardness test is the most widely used of all hardness tests, and it occupies such an established position that the Brinell hardness number is a generally accepted measure of hardness. The Brinell hardness number as defined by the British Standard Specification No. 240-1937 is the quotient of the applied load divided by the spherical area of the impression and is given by the formula—

$$H = \frac{P}{\frac{\pi D}{2}\left(D - \sqrt{D^2 - d^2}\right)} = \left(\frac{P}{D^2}\right)\left(\frac{2/\pi}{1 - \sqrt{1 - (d/D)^2}}\right)$$

where P is the load in kilogrammes, D is the diameter of the ball in millimetres, d the diameter of the impression in millimetres, and H the Brinell hardness number.

Some 20 to 25 years ago the Brinell hardness test was more or less firmly established, but hardness tests were limited to those made with the 3,000 kilog. machine. By removing weights and reducing the size of the ball, relatively low load tests could be made, but when the hardness of thin material had to be developed only the Shore scleroscope was available. Since then a wide variety of Brinell hardness type testing machines have been developed in which considerable attention has been paid to the magnitude of the load, to the accuracy of testing, to measuring devices, and to the size and type of the ball used.

A large number of investigations have been carried out on the Brinell hardness test, and in his book on the "Hardness of Metals and its Measurement," O'Neill[1] not only deals with these investigations, but also

1 H. O'Neill. "The Hardness of Metals and Its Measurement," 1934, Chapman and Hall, Ltd.

summarises in a very comprehensive manner the principles underlying this test. This author also gives consideration to the large amount of test data which has been accumulated, and to the manner in which it is related to the physical properties of the materials tested, and also discusses the characteristics of the balls used and of the procedure adopted in testing.

A more recent study of the Brinell hardness test and of the effects of small variations in testing procedure, which may influence observed readings, has been made by Petrenko, Ramberg and Wilson.[2] To supplement existing knowledge, tests were carried out on 37 different metals and alloys to account for and remove such discrepancies in measurement of Brinell hardness as may occur due to the apparatus and procedure, to non-uniformity of specimen, to the curvature and thickness of specimens, to the spacing of indentations, to the inclination of load to the normal and to the shape and size of the ball and its deformation under load.

As a result of this investigation, it is recommended that the ball should be loaded at a rate not exceeding 500 kilogs. per sec., and that the load should be applied for 30 secs. The error in the load should not exceed 0·5%, and this should be checked by periodic calibration, as should also the microscope or apparatus used for measuring the diameter of the indentation. The average of four equally spaced readings should be taken unless the specimen is curved, in which case the two principal diameters should be averaged. The test specimen should be at least 0·4 in. thick, the angle between the load line and the normal to the specimen should not deviate more than 2°, and the distance between centres of adjacent indentations should be at least three times the diameter of the specimen. Petrenko, Ramberg and Wilson also investigated the correlation necessary for the elastic deformation and the permanent set of steel and tungsten carbide balls (Fig. 1), and recommended that a difference between the average diameter and the normal diameter (10 mm.) of the ball should not exceed 0·25 mm. (0·001 in.). The material of the indenting ball should be specified in recording Brinell hardness number greater than 500, and the permanent compression of the loaded diameter of the ball after an indentation of a specimen having a Brinell number less than 500 should not exceed 0·001 mm. The use of carbide tungsten balls is recommended for indentations on any specimens having a Brinell hardness number greater than 500, and if steel balls are used on such specimens the permanent compression after any indentation should not exceed 0·025 mm.

It is of interest to compare the conclusions of these investigators with the requirements of the British Standard Specification for Brinell Hardness Testing No. 240-1937. In this specification the accuracy of testing is required to be within 0·5% of the load applied to the test specimen. The micrometer microscope or device used should be capable of measuring the diameter of the impression to within ± 0·5%. It is also specified that the balls used shall be of hardened steel or some harder material and should not differ from the appropriate standard diameter by more than ± 0·0025 mm. (0·0001 in.). The thickness of the test specimen should be at least 10 times the depth of the impression, and the centre of the impression not less than two and a half times the diameter of the impression from any edge of the

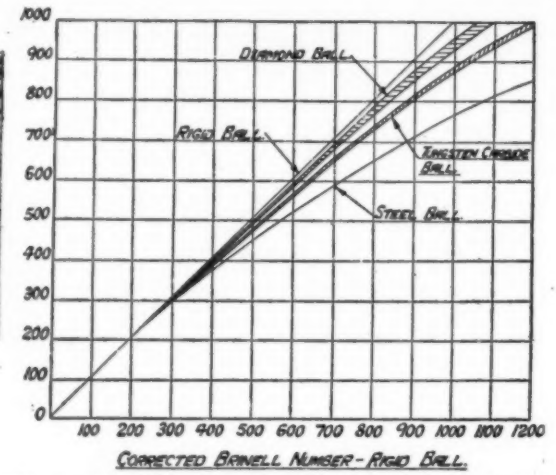


Fig. 1.—Brinell hardness as determined with steel, tungsten carbide, and diamond ball, compared with corrected value for a perfectly rigid ball. (Petrenko, Ramberg and Wilson).

specimen. Where balls of 1 mm. or 2 mm. diameter are used, the test surface should be polished, and with 5 mm. or 10 mm. diameter balls a surface finish obtained by filing, grinding, or smooth maching is adequate. The specification also states that the test shall be carried out with loads and ball diameters selected from Table I, corresponding to the value of the ratio $\frac{P}{D^2}$ stated in the specification for the material under test, and that the load shall be applied slowly and progressively and should be maintained for 15 secs.

TABLE I.
RELATION OF BALL TO LOAD IN BRINELL HARDNESS TESTING
(BRITISH STANDARD SPECIFICATION No. 240-1937).

Diameter of Ball, Mm.	$\frac{P}{D^2} = 1$ Kilog.	$\frac{P}{D^2} = 5$ Kilog.	$\frac{P}{D^2} = 10$ Kilog.	$\frac{P}{D^2} = 30$ Kilog.
1	1	5	10	30
2	4	20	40	120
5	25	125	250	750
10	100	500	1,000	3,000

With steel the resistance to deformation indicated by the Brinell hardness number corresponds to the tensile strength, and is therefore a convenient means for evaluating this property. With non-ferrous metals, however, this simple relationship does not exist, as two different alloys having the same tensile strength may have two different Brinell hardness values. It has been shown, however, that in such cases the metal with the lower elastic limit has the lower Brinell hardness number, and it is therefore possible within limits to arrange non-ferrous metals in order of increasing strength by means of their Brinell hardness numbers. With cast iron varying results are obtained. A recent German investigation carried out by the Cast Iron Sub-Committee of the Verein deutscher Eisenhüttenleute[3] on the spread of Brinell hardness determinations on cast iron, using a 10 mm. ball and a 3,000-kilogs. load, has shown that no relation can be established between the spread and the hardness in the range of 120 to 180 Brinell. When the hardness increased to 250 the structure affected the results and a marked change was found in the spread.

2 N. Petrenko, W. Ramberg, and B. Wilson. *J. of Research*, Nat. Bureau of Stand., 1936, vol. 17, pp. 59–95.

3 E. Pohl and H. Eisenwiener. *Arch. für das Eisenhüttenwesen*, 1941, vol. 14, pp. 391–396.

The spread, which was ± 13 Brinell units, when the mean reading was 120, was reduced to ± 5 or 6 units for a mean reading of 250. It was also shown that the spread of hardness determinations on pearlitic cast iron was not dependent on the matrix, but on the quantity and form of the graphite.

Diamond Pyramid Hardness Test

Diamond pyramid hardness tests are based on the same principle as ball indentation tests—namely, that the dimensions of the impression made in the material by the point of a pyramid or prism-shaped indenter under the action of a standard load are used to determine the hardness of the material. The diamond pyramid hardness number as defined by the British Standard Specification No. 427-1931 is the quotient of the applied load divided by the pyramidal area of the impression and is given by the formula—

$$H_D = \frac{2\ P\ \mathrm{Sin}\frac{\theta}{2}}{d^2}$$

where P is the load in kilogrammes, d the diagonal of the impression in mm., θ the angle between each pair of opposite faces of the pyramidal diamond, and H_D the diamond pyramid hardness number.

The pyramid indentation method of hardness testing has been developed to overcome certain disadvantages associated with the ball method of testing. In ball hardness testing, limitations result due to ball deformation on very hard surfaces so that even with the hardest type of balls the upper limit to which they can be used is 700 on the Brinell scale. In ball testing there is also the destruction of an appreciable area of the test specimen, particularly in soft material. In the pyramid hardness test, as a result of the lighter loads applied and the smaller size of the indenter, the impressions made on a surface are so small that this method of hardness testing can be used for certain classes of finished parts and so becomes a non-destructive test. With pyramid-shaped indenters impressions are similar, no matter what their size, so that lack of similarity between impressions of different diameters, which may occur with ball indenters, do not exist. Most types of pyramid hardness testing machines are equipped with indenters made from diamonds ground to pyramid size and mounted in a metal cup. The best known types of pyramid hardness testing instruments are the Vickers' pyramid hardness testing machine and the Firth hardometer.

In the British Standard Specification No. 427-1931, for diamond pyramid hardness testing, it is stated that the indenter should be in the form of a square pyramid and have an angle between opposite faces of 136°. Where a diamond having an angle between opposite faces of 140° is used, the resulting hardness numbers will be lower than those obtained under standard conditions by 2 to 3·5%. The micrometer, microscope or other measuring device used to measure the impression should be capable of measuring each diagonal of the impression to an accuracy of ± 0·5% (± 0·1001 mm.), and the standard loads used for testing are specified to be 5, 10, 20, 30, 50, 100 and 120 kilogs. For soft materials or coarse-grained materials, particularly those having a duplex structure, the indenter may be made from hardened steel or some hardened material having a diamond pyramid hardness number 400 units greater than the material under test, and in such cases loads of 500 and 3,000 kilogs. may be used. In testing, the centre of the impression should not be less than two and a half times the diagonal of the impression from any edge of the test specimen, and the thickness of the test specimen should be at least equal to one and a half times the diagonal of the impression. Loads should be applied slowly and progressively and maintained for 15 secs.

Rockwell Hardness Test

The Rockwell hardness test also includes a diamond indenter ground to the form of a cone with an included angle of 120°. A hardened steel ball may also be used. In this test (Fig. 2) a minor load P_0 is applied to the penetrator and causes an indentation in the test specimen. While this load is still in operation, it is agumented by a major load, P_1, with resulting increase in the depth

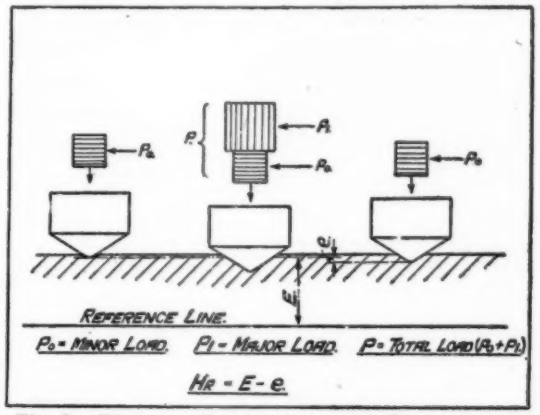


Fig. 2.—Direct reading hardness test, Rockwell principle. (British Standard Specification No. 891-1940).

of indentation. The major load is then removed, still retaining the minor load, and this operation effects a partial recovery in the depth of indentation and eliminates the indication of the deflection in the structural members during application of the major load. Rockwell hardness numbers are generally read directly on an indicator, usually a sensitive dial-type depth gauge calibrated to read hardness numbers directly, and are derived from the measurement of the depth of the impression according to the equation—

$$H_R = E - e$$

where e is the difference between the depth of penetration before and after the application of the major load and while the minor load is in operation in both cases, E an arbitrary constant depending on the form of the penetrator used, and H_R the Rockwell hardness number. The value of e is expressed in units of 0·002 mm.

The Rockwell hardness test is useful for repetition work on certain simple-shaped parts, but it has sometimes been found that results for this test obtained with different testing machines give results not in accordance with each other. The adoption recently of a British Standard Specification No. 891-1940, dealing with direct hardness testing, Rockwell principle, is of value in that it deals not only with the apparatus used and the test specimen, but also deals with the use of scales A, B and C, additional scales, and a method for checking the accuracy of the test.

In this specification it is stated that the conical penetrator shall be a diamond with an angle of 120° to

an accuracy of ± 0·1° in any axial plane. The point of the penetrator shall be rounded to a radius of 0·20 mm., and the profile shall conform to the radius of 0·20mm. within an accuracy of ± 0·005 mm., and the round and conical surfaces shall be joined in a truly tangential manner. Where a ball penetrator is used it shall comply with the requirements of British Standard Specification 240, Part 2, on the hardness of steel balls for Brinell testing, and shall be of a diameter of $\frac{1}{16}$ in. (1·588 mm.) and its diameter at any part shall not differ from the standard diameter by more than ± 0·0001 in. (0·0025 mm.).

The standard Rockwell test should be carried out in accordance with one or other of the scales A, B and C, as given in Table II, and the accuracy of the applied loads should be within ± 0·5%. In measuring the depth of penetration, the measuring device should be capable of measuring the depth of the impression to within ± 0·001, and a sliding scale should be used which can be set at zero following the first application of the minor load, so as to enable the difference in depth before and after application of the major load to be measured directly by a single reading.

TABLE II.
ROCKWELL HARDNESS SCALES A, B, C.
BRITISH STANDARD SPECIFICATION No. 891-1940).

Scale.	Penetrator.	Minor Load. Kilogs.	Major Load. Kilogs.	Total Load. Kilogs.	Value of E.
A	Diamond cone	10	50	60	100
B	Steel ball, $\frac{1}{16}$ in. (1·588 mm.)	10	90	100	130
C	Diamond cone	10	140	150	100

The surface on which a Rockwell impression is made should be flat, as test results obtained on curved surfaces are only of comparative values. The thickness of test specimens should be sufficient to avoid any bulge on the surface opposite the impression and resulting from the application of the loads, and the centre of the impression should not be less than two and a half times the diameter of the impression from any edge of the specimen or from any other impression. It is also necessary that the test specimens, in whatever form, should be rigidly supported, and that the framework of the testing apparatus or machine should be sufficiently rigid to prevent the possibility of strains in its members affecting the magnitude or duration of the loads. The same also applies to the depth-measuring device.

In carrying out Rockwell hardness tests, scale A, using a diamond cone and a total load of 60 kilogs. is recommended for thin hardened steel strip and other extremely hard materials where small impressions are required. For all mild and medium carbon steels, sheet steel and soft steel bars, Scale B, using a $\frac{1}{16}$ in. steel ball and a load of 100 kilogs., is most suitable, while hardened steels, hardened and tempered steels and alloy steels should be determined on Scale C, using a diamond cone and a total load of 150 kilogs. For determining the hardness of the very thin case produced by the nitriding process, shallow case-hardening, decarburisation, surface cold-working or by other means, Scales A, B and C are not suitable. Special tests based on the same principle and using very light loads have to be made available. Additional scales such as are given in Table III, are also used for special purposes.

With a view to correlating readings made with different loads and ball diameter in the Rockwell hardness test, and in an endeavour to simplify the relation between hardness and tensile strength measurements, Peek and Ingerson[4] have experimented with test specimens of various materials, including aluminium, mild steel, six brasses, four nickel silvers, three phosphor bronzes, and five thicknesses of half-hard grade of nickel silver. These various alloys were tested with different loads and ball diameters, and their tensile strength was subsequently determined. The results obtained not only showed a good relationship between the different tests, but also indicated the possibility of calibrating the Rockwell hardness tester with a single test block, using various loads and ball diameters, and of setting up specification limits for hardness measurements corresponding to tensile strength limits.

TABLE III.
ROCKWELL HARDNESS ADDITIONAL SCALES
(BRITISH STANDARD SPECIFICATION No. 891-1940).

Scale.	Penetrator.	Minor Load. Kilogs.	Major Load. Kilogs.	Total Load. Kilogs.	Value of E.
D	Diamond cone	10	90	100	100
E	Steel ball, $\frac{1}{8}$ in. (3·175 mm.)	10	90	100	130
F	Steel ball, $\frac{1}{16}$ in. (1·588 mm.)	10	50	60	130
G	Steel ball, $\frac{1}{16}$ in. (1·588 mm.)	10	140	150	130
H	Steel ball, $\frac{1}{8}$ in. (3·175 mm.)	10	50	60	130

Monotron Hardness Test

The Monotron hardness test is a static hardness test, using as an indentor a sphere with a diameter 0·75 in. It differs in principle from other hardness static tests, in that it measures the loads required to make an arbitrary impression 0·045 mm. deep, 0·33 mm. diameter. The load in kilogrammes required to produce this impression is a measure of the hardness. Instruments used for determining the hardness by this test are usually fitted with two dials or gauges, one which registers the depth of the impression whereby an impression of predetermined depth may be obtained, and the other which measures the load to produce the given impression. The Monotron hardness test differs from the Brinell hardness test in that the deformation produced by the indenter includes the elastic as well as the plastic deflection.

The Monotron hardness test can also be used to determine hardness by determining the depth of penetration for any given constant load. In this respect it is used in measuring the depth of hardness, as in a material of uniform hardness the load and depth of penetration follows practically a straight line, so that by plotting values obtained for different penetrations, variations of hardness are indicated by any deviations which may occur.

Shore Scleroscope

In the rebound method of hardness testing, the Shore scleroscope is the test which has been most extensively used. The principle employed in this method of testing is the drop and rebound of a diamond-tipped hammer about one-twelfth of an ounce in weight. This hammer falls from a height of 10 in. on to the test specimen, and on striking the surface of the specimen makes a small indent or penetration, the size of which depends on the material tested, and then rebounds. As the work of deformation absorbs some of the energy of the fall,

4 R. L. Peek and W. E. Ingerson, Amer. Soc. Test. Mat. Preprint., No. 106, 1939, 10 pp.

the harder the material tested the less energy absorbed and the greater the rebound, and softer the material tested the more energy absorbed and the less the rebound. The height of the rebound is measured against a graduated scale on which quenched high carbon steel gives a rebound of 100.

In carrying out the Shore hardness test, it is essential that the instrument be kept perfectly vertical, that the glass tube be kept clean and free from dirt and grease, and that the test specimen is smoothly polished, perfectly horizontal, and is rigidly held. The supporting or clamping of the test specimen is most important, as its mass or inertia has a most important effect on the rebound. Experiments have shown that as the mass decreases sufficiently to cause a falling off in rebound, the softer metals show that greater discrepancy due to the deeper penetration and the greater time factor which takes place, making the inertia of the mass less effective against the blow of the hammer. Special methods of clamping and support must, therefore, be used to provide the mass and inertia that sheets, flats and other underweight parts lack, if accurate results are to be obtained.

The rebound method of hardness testing is particularly suitable for the testing of small hardened steel parts, but can also be used for any class of steel and for non-ferrous metals. When testing the latter or testing soft steels, a steel-tipped hammer, known as a magnifier hammer, is used in place of the diamond-tipped hammer. As this hammer has a rounded striking point, there is less penetration into soft materials, and consequently a rebound to a greater height than if a diamond-tipped hammer is used. A special scale of hardness numbers has to be used with this hammer.

There are two forms of Shore scleroscope : the ordinary type and the recorder type. In the latter, a heavier hammer is used and the height of fall reduced so that the same energy of fall is attained. On account of the different weights of the two hammers, the size of the mass tested has a greater effect with the heavier hammer than with the lighter hammer, and in some cases results are obtained which are not in agreement, the recording type of instrument giving the higher value. On this account the type of instrument used has to be considered when testing for specification purposes.

Herbert Pendulum Hardness Tester

The Herbert pendulum hardness tester is a simple type of instrument which is very easy to use. The principle on which this test is based differs from other types of hardness tests. In the pendulum hardness tester, use is made of the alteration of the time of swing of the pendulum by change of the centre of oscillation when a spherical-shaped diamond indentor, which is the point of support of the pendulum, moves in the groove made by itself. With this type of hardness tester the hardness of all metals and alloys can be determined, no matter how hard they are. It is also suitable for testing small test specimens and thin materials.

While this instrument has not been used to any great extent for industrial hardness testing, it has been found suitable for experimental and research work. With it investigations have been carried out on the work-hardening capacity of metals, and on the super-hardening capacity of hard steels. As it is suitable for use with hot-test specimens as with cold-test specimens, it has been found of value in determining the hot hardness of tool steels and heat-resisting alloys. It has also been used in determining low-temperature changes in metals and alloys.

A continuous hardness test developed by Herbert[5] from the pendulum hardness test has been found suitable for recording hardness changes occurring in metals during ageing. In this test a loaded ball free to rotate is run over the test specimen, and with the aid of an optical lever hardness is recorded as a fraction of time. Tests with this instrument on an aluminium alloy undergoing precipitation hardening showed periodic fluctuations to be superimposed on the general increase in hardness brought about by precipitation. With another instrument periodic fluctuations were also found in hardened steels when a diamond with a hemispherical tip was loaded and propelled over the test specimen.

Hot Hardness Testing

Hot-hardness tests have been carried out, using both static and dynamic methods. At elevated temperatures difficulties are met with when using static testing methods, and these become more marked the higher the temperature. These difficulties include oxidation, flattening and structural changes of the ball or indenter, and continuous flow of the material under load. Preheating the indenter, the use of indenters made of tungsten carbide or single crystals of corrundum, and testing in inert gases have eliminated some of the difficulties, but the plastic flow of the metal being tested remains. This factor is absent, however, in dynamic hardness testing. An interesting review of high-temperature dynamic hardness testing methods has been made recently by Fetz,[6] who discusses the fundamental principles of dynamic hardness testing, the design and operation of typical hardness testing machines, and also gives a compilation of data obtained in various alloys by hot dynamic hardness tests.

Fetz also reports the results of experiments in which dynamic hardness tests were used to investigate the softening of die steels, the influence of silicon additions on the properties of 18/8 chrome-nickel steels at hot-working and at service temperatures, the influence of molybdenum on the high-temperature hardness of 12% chromium steels and the influence of rising temperature on the resistance of 25/19 nickel-chrome steel to plastic deformation. A guillotine type of hardness testing instrument was used for the investigation.

Hardness Conversion Scales

One of the difficulties encountered by users of hardness testing equipment is the need for accurate comparison of the units of hardness obtained by the various methods, a need which has been increased during the last 10 to 15 years by the increasing adoption of hardness testing as an inspection method. The most widely used method of measuring hardness are the static penetration tests represented by the Brinell, Pyramid Hardness and Rockwell tests, and while there are individual preferences for using either of these methods, cases are increasing where the same material is tested in different establishments by the three different methods, with the result that there has been a demand for some means of correlating the various methods of hardness testing. Work has therefore been carried out during recent years, both in Britain and the United States, with a view to providing accurate hardness conversion tables.

5 E. G. Herbert, METALLURGIA, 1937, vol. 16, pp. 184–186.
6 E. Fetz. Amer. Soc. for Metals, 1941, Preprint 38, 39 pp.

In Britain, a table of approximate hardness scales has been prepared by the Mechanical Industry Committee of the British Standards Institution and issued as a British Standard Specification No. 860-1939. The Diamond Pyramid Scale is taken as the basis of reference, and the most probable comparative values, Brinell and Rockwell A, B and C Scales have been adopted from published experimental results. Some of the comparison values from this specification are given in Table IV. In connection with the various scales, investigations show that there is no general theoretical relationship between these scales and that empirical formulæ devised from experiments only hold closely for materials of approximately similar composition and in a given condition. If the conditions as regards composition, heat-treatment, and mechanical treatment differ appreciably, and if in the Brinell hardness test different loading ratios are used, variations occur in the empirical relationships. On the other hand, groups of materials similar as regards composition and other conditions, when tested by any of these penetration methods, give fairly close comparisons. The table of hardness comparison scales contained in the specification is a general approximation and is issued only as an indication of the order of relationship between the three methods of determining hardness, and must not be used as a standard for the conversion of hardness values given on one scale in any British Standard Specification to those of another scale.

TABLE IV.
APPROXIMATE COMPARISON OF HARDNESS SCALES
(BRITISH STANDARD SPECIFICATION No. 860-1939).

Diamond Pyramid Scale (B.S. 427-1931).	Brinell Scale (B.S. 240-1937).	Rockwell Hardness Scales (B.S. 891-1940).		
		A Scale 60-kilog. Diamond Cone.	B Scale 100-kilog. $\frac{1}{16}$ in. Steel Ball.	C Scale 150-kilog. Diamond Cone.
20	20	—	—	—
100	95	43	54	—
200	195	59	94	—
300	295	66	—	30
400	380	71	—	40
500	460	75	—	48
600	535	77	—	54
700	595	80	—	58
750	630	81	—	61
800	—	82	—	62
1,000	—	84	—	68
1,200	—	87	—	70
1,400	—	91	—	71

In the United States tests on the relationship between the various methods of hardness testing have been carried out under the auspices of the Inspection and Standards Department of the Westinghouse Electric and Manufacturing Co., and two reports on the work have been published. The first report[7] deals with the relationship between the Diamond Pyramid hardness and Rockwell C hardness, and the second report by Scott and Gray[8] covers experiments on hardness conversion relations for hardened steel and other alloys. Using the Diamond Pyramid hardness test as a reference standard, a conversion table has been prepared for hardened steel and other alloys which permits conversion from a hardness value obtained by either the Diamond Pyramid, Rockwell Scales, Monotron, Scleroscope and Brinell hardness tests to any other.

Tests carried out in preparing this table on steels and sintered carbides showed that the relation between the Diamond Pyramid hardness and the Rockwell hardness was dependent on the modulus of elasticity of the materials tested, so that a separate table is necessary for metals and alloys differing materially in elastic modulus from steel. The relation between Rockwell and Diamond Pyramid hardness test values on steels is expressed accurately by a simple empirical formula, and with the change of a single constituent the same equation gives equally well the same relation for sintered carbides. Adding a term to this equation for elastic modulus, new constants for other alloys were evaluated from the determinations on steel and sintered carbides, and values calculated for metals and alloys of low elastic modulus, such as copper and aluminium alloys, agree well with experimental determinations.

A special Committee of the American Society for Testing Materials[9] has also had the subject of hardness conversion tables under review for some time, and in view of the urgent need for such information on cartridge brass, hardness conversion tables for this alloy have been dealt with as a first stage of the investigation. A co-operative research has been carried out in a number of laboratories in each of which blocks of various tempers of brass have been subjected to hardness tests, using the Diamond Pyramid hardness test with loads of 10, 20 and 30 kilogs., and the Rockwell hardness tests B and F scales and certain additional scales. Brinell hardness tests were also made on blocks which were sufficiently hard to obviate the anvil effect. The data obtained from the results of these tests, together with data already available at the respective laboratories, have been correlated and published in the form of a Proposed Conversion Table for 70 : 30 brass.

From the work which has already been carried out, and from the conversion tables prepared, it would appear, therefore, that there is sufficient evidence available for the use of a single reference scale of hardness for determining all penetration hardness values, irrespective of the type of test used.

Testing Cylindrical Parts

Another difficulty met with in hardness testing is that experienced in the testing of cylindrical or round parts. As the direct hardness measurement of such surfaces is difficult and unreliable, theoretical methods of calculating the hardness are adopted. One such method for correcting the apparent Rockwell hardness values obtained on cylindrical specimens using the ball indenter has recently been developed by Ingerson,[10] who calculated the deviation of apparent from true hardness for various specimens from $\frac{1}{8}$ in. to 1 in. diameter, and found it to vary approximately linearly with the depth of impression for a given test specimen. Tests were made on steel, brass, aluminium, phosphor bronze, and nickel silver, using the Rockwell B, F and G scales. The variation in apparent hardness measured on flat and curved surfaces was found to check with their differences to within 0·5% of a Rockwell unit. Correction curves were derived from these tests, and by the use of these curves true Rockwell hardness values on the B, F and G scales can be obtained from measurements made on cylindrical surfaces.

In a recent Russian investigation, Spektor[11] also derives a formula for determining the hardness of spherical surfaces by the Rockwell method. Experiments

7 E. Fetz, *Trans. Amer. Soc. for Metals*, 1939, vol. 27, p. 363.
8 H. Scott and T. H. Gray, *Trans. Amer. Soc. for Metals*, Preprint No. 28, 1939, 18 pp.

9 J. R. Townsend, *Bull. Amer. Soc. Test. Mats.*, 1942, vol. 115, pp. 35–36.
10 W. E. Ingerson, Amer. Soc. Test. Mats., Preprint No. 106, 1939, 10 pp.
11 A. G. Spektor, "Zavodskaya Laboratoriya," 1940, No. 4, pp. 455–460 (*Jour. Iron and Steel Inst.*, 1942, vol. 145, p. 153A).

on steel balls showed that the calculated results were correct up to a radius of curvature of 6·5 mm., but for larger radius of curvature values 6·5 mm. to 11 mm., and 11 mm. to 15 mm. the error amounted to 0·5 to 1 Rockwell C units respectively.

Micro-Indentation Hardness Testing

A third difficulty in hardness testing is that met with in testing small specimens and thin sections, and for such parts micro-indentation methods have been developed principally in Germany and the United States. During recent years such methods of testing have made considerable progress and have been found useful in examining metal foils and very thin plated and treated surfaces, as well as individual micro-constituents, enamels, ceramics, and minerals.

The specialised methods used in these tests have recently been reviewed by O'Neill,[12] who has summarised the various instruments available for this type of test, giving details of the type of indenter used, the load required, and the principle involved. Two instruments, the Zeiss micro-indenter and the Knoop micro-indenter, are briefly described. A comparison is made with the results obtained with the latter instrument and with two other micro-indenters on such micro-constituents as silicon carbide, carborundum, martensite and tungsten molybdenum carbide. Data is also given for the hardness of the constituents or iron and steel and the constituents of tin-base, lead-base, cadmium-base, and copper-lead bearing metals. Hardness values are also given for the outer layers of metals treated by carburising and quenching, cyaniding and quenching, nitriding, cementation, sheardising, chromium plating, etc.

The Knoop micro-indenter is also described in detail by Peters and Knoop,[13] and a series of hardness tests on a number of metallurgical materials are given to show its applications. Some of these determinations were made on the hardness of the nitrided skin on molybdenum high-speed steels, with different times and temperatures of salt-bath treatments, the hardness of large carbide particles in high-speed and plain carbon steels, the hardness of sintered carbides, including boron carbide, and the hardness of a number of other materials including chromium plate and phosphor-bronze sheet. Indentations were also determined for standard Rockwell specimens and for rolled copper, both with this micro-indenter and with the Brinell and Vickers' hardness machines. The results obtained are given in Table V. The data from these tests show that the micro-indentation number obtained for rolled copper is 20% greater than the Brinell hardness number and 30%

TABLE V.
COMPARISON OF MICRO-IDENTATION HARDNESS WITH BRINELL AND VICKERS' HARDNESS FOR THE SAME SPECIMENS (PETERS AND KNOOD).

Specimen.	Ball.	Load (30-sec.) Kilogs.	Brinell Hardness.	Vickers Hardness.	Indentation Hardness.
Rockwell C, 65–67	10 mm. carboloy	3,000	780	894	791
Rockwell C, 47 ...	10 mm. steel....	3,000	456	493	496
Rockwell C, 25 ...	10 mm. steel ...	3,000	241	265	276
Rolled copper	$\frac{1}{8}$ in. steel	15	104	111	125
Rockwell B, 7–9 ..	10 mm. steel....	500	52·8	58·2	59·2

greater than the corresponding Vickers' number, while the micro-indentation number for steel C 65–67 agrees with the Brinell hardness number, but is 12% less than the Vickers' hardness number for the same steel.

12 H. O'Neill. METALLURGIA, 1941, vol. 23, pp. 71–74.
13 C. G. Peters and F. Knoop. *Metals and Alloys*, 1940, vol. 12, pp. 292–297.

Scratch tests and the various forms of micro-indentation tests, while useful within limits, are not entirely satisfactory for all cases in which hardness values of thin layers are required, layers such as hard, electro-plated white bronze castings and polished or anodised surfaces. A new method developed by Chalmers[14] for testing thin electro deposits has been found satisfactory and is of much wider potential value in that it can be used for determining the hardness of oxide films, polished copper and tin surfaces, the surface layers of heat-treated tin and copper, and the cathodically treated surface layers of tin, nickel, white bronze, silver and copper.

This new method of testing is based on the principle that while it is difficult to apply a small indenting force under accurately standardised conditions, it is possible to apply a very large number of such forces in a way that while each is under control the result is statistically definite. The procedure is to drop a large number of grains of sand, or particles of emery on to the specimen. The particles are of varying shapes, and within limits fixed by the screening of the sand of varying size. They are, in addition, randomly orientated when they hit the surface. The impressions in the surfaces, therefore, vary widely in shape and size, but if a sufficiently large number of particles are used, the average shape and size of the pits depend on the amount of sand used. It is therefore necessary to measure the average size of the indentations, a result which is arrived at by measuring the reduction in the specular reflectivity of the surface.

The results obtained by this method of testing show that while the method is not applicable for measuring bulk hardness of metals and alloys, it is capable of providing useful data on the hardness of surfaces prepared in various ways, the only condition being that the surface should possess a fair degree of specular reflection. Data obtained on various non-ferrous metals and alloys with a view to relating the hardness obtained by this method with the Vickers' Diamond Pyramid hardness show a general relationship to exist between these two different methods of determining hardness.

Substitute Materials

IN normal times the industrial countries draw upon the world for their supplies of raw materials, but when conditions arise which interfere and ultimately prevent the free exchange of goods or materials, shortages occur, and those countries directly affected must endeavour to make good the shortage by substituting other materials which are available. This subject is admirably discussed by Dr. C. H. Desch in a booklet recently published by the Royal Institute of International Affairs.

The experience of 1914–18 and the developments in the inter-war period are considered in the introductory section, while the remaining section is concerned with the conditions arising as a result of the present conflict. In this latter section are discussed synthetic substitutes for petroleum, substitutes for rubber, plastics, other organic materials and metals, while consideration is also given to the economical use of scarce materials. It is obvious that the development of many materials will have a considerable influence on industry after the war, and in this booklet will be found useful.

It is published by the Royal Institute, Chatham House, St. James's Square, London, S.W. 1 ; price 9d. net.

14 B. Chalmers. *J. Inst. of Metals*, 1941, vol. 67, pp. 295–314.

The Manufacture of Non-Ferrous Seamless Tubes

By Gilbert Evans

Extrusion practice and plant are showing progressive tendencies. Improvements in plant are permitting the operations to be carried out at greater speed and preparatory operations formerly considered necessary are being eliminated without detriment to the finished tube. Attention is directed more especially to condenser tubes, the production of which is making special demands upon the trade in these times.

THE demand for war material has resulted in vast expansion and additions to plant in all branches of tube manufacture; in particular, the rapid growth of the merchant navy and of vessels engaged in patrol and convoy work has greatly increased the need for condenser tubes, and manufacturers have been, and continue to work, under high pressure to meet the abnormal demands necessitated by war conditions. Fortunately, manufacturing developments incorporated in the last decade have greatly facilitated the speed of production, as well as effecting improvements in the finished tube.

About twenty years ago an expert manufacturer stated: "As long as there is a demand for non-ferrous tubes, the casting of hollow shells on sand cores will survive." While this method of production persists in some of the smaller works, the development of vertical and horizontal extrusion press practice has eliminated casting on sand cores in progressive tube works. This is one of several factors that has contributed to increased production.

The writer has had working contact with the majority of types of plants engaged in the production of this class of tube, including the Schloeman, Krupp, and Hydraulik plants emanating from Germany, and particularly Fielding and Platt's, many of the latter plants being installed in a number of British tube works. As an indication of compactness and simplicity, it is noteworthy that in a recent description of the Krupp extrusion press reference was made to the press unit comprising nineteen parts, whereas, by comparison, the press illustrated in Fig. 1 is covered by less than half that number. In both instances neither pumps nor accumulators are included in the number of parts mentioned. Modifications in design have resulted in greater simplicity, which have proved effective in increasing productive capacity.

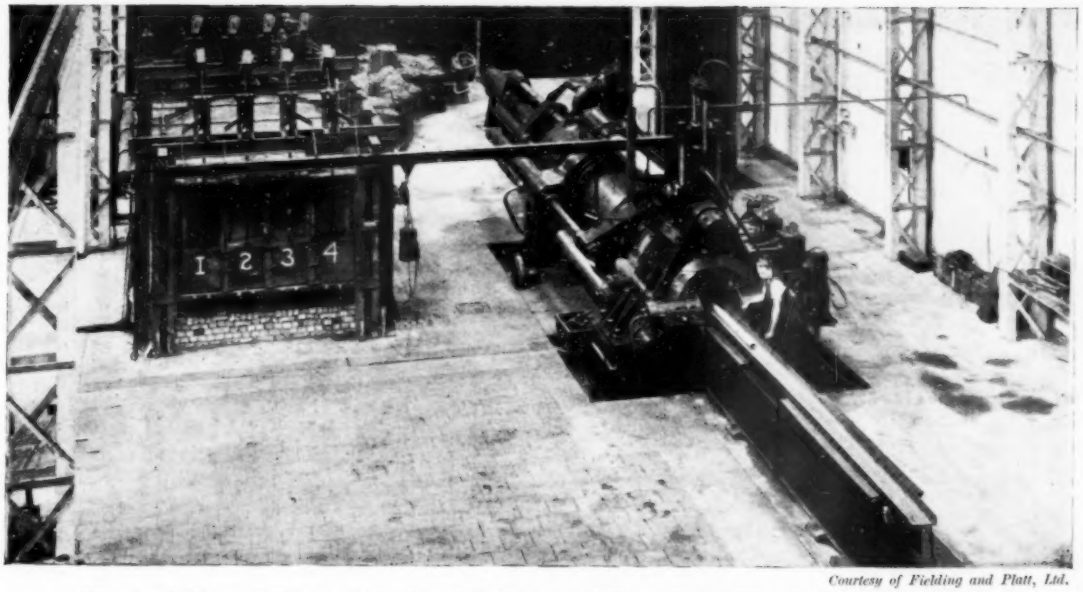
In view of the progress achieved in this branch of the non-ferrous trade, it will be of interest to describe briefly the various operations involved in the production of condenser tubes, which consist essentially of the melting of the metal and subsequent casting into billets: heating the billets and extruding them to the required tube sizes, using a 70-copper and 30-zinc composition which is familiar in the trade.

The charge of copper and zinc, together with suitable scrap, is carefully weighed, making suitable allowance for loss of zinc through volatilisation—in a well-operated foundry this loss should be low. The copper is melted first and the zinc added gradually, protecting the surface of the metal against contact with air. Crucible or everberatory furnaces may be used, but in most modern tube works electric furnaces of suitable capacity are usual. The metal should be melted under strict pyrometer control, so that casting can be effected at an approved temperature.

The moulds into which the metal is cast usually conforms to one of two types, either water-cooled or cannon-shaped moulds. The former consists of a water-jacketed copper tube corresponding to the diameter of the billet; the water being circulated during the time of casting. The cannon-shaped mould consists in the main of a single casting, in dense grey iron, which is carefully bored to the required diameter of the billet. The casting surface must be clean and free from imperfections. As a rule, the mould is parallel, although some manufacturers prefer a slight taper to facilitate stripping, in which case the smaller end is cast uppermost. The surfaces of all moulds must be clean, and they are usually dressed with a mixture consisting mainly of animal fat, china clay and graphite. This dressing is used to improve the surface of the billet and to assist its extraction after the metal has solidified, thus the surface of the mould must be cleaned and dressed for each casting. In most modern shops the moulds are mounted on rotary tables, and casting is continuous, succeeding moulds being brought directly under the pouring spout of the ladle or furnace as the table is rotated.

On extraction from its mould, this billet is frequently surface skimmed in a lathe and examined for defects; in many cases, however, this skimming operation is considered unnecessary. The billets are cold sawn to lengths which are governed by the length of the tube product; these cut lengths are often bored with a $1\frac{3}{4}$ in. special drill, though in some works drilling is not considered necessary. The main object in boring the billets is to assist in obtaining a concentric product from the extruding operation, but it also reduces the pressure on the mandrel bar when piercing.

The billets are reheated preparatory to being extruded, and many types of furnaces are designed for this purpose which may be heated by solid fuel, fuel-oil, gas, or electricity. The billet-heating furnace shown in Fig. 1 is heated by fuel-oil. It is approximately 20 ft. long by 9 ft. wide, and comprises four sections with independently operated charging and discharging doors. The bottom of the heating chamber is arranged for gravity discharge, and is sloped to 2 ft. 6 in. above floor level at the discharging end. This height coincides with the working centre of the horizontal press. The billets are fed to the heating chamber from a lifting table. Heating of the billets is arranged to coincide with the time required for extrusion and to maintain continuous operation. An H-girder is attached to the furnace to carry a



Courtesy of Fielding and Platt, Ltd.

Fig. 1.—Main extrusion plant for non-ferrous tube manufacture, including a 1,500-ton horizontal hydraulic extrusion press and an oil-fired billet reheating furnace.

runway, so that the billets are conveniently transported to the press.

Extrusion Operation

It is proposed to describe an outline of the operations in the various stages of production of standard non-ferrous shell to be cold-drawn to condenser tube size, as required to meet condenser tube specification. The average size of billet used is $6\frac{7}{8}$ in. diameter and $10\frac{1}{2}$ in. long, which is extruded to a hollow section having an outside diameter of 3 in., a $2\frac{3}{8}$ in. bore, and averaging 8 ft. in length. The billet is heated to a temperature of 780°–820° C. in the furnace and is transported to the container of the press, where it is brought into contact with the extrusion die or matrix. The press ram is retracted and the guide bush and pressure plate inserted in the container. The piercing mandrel is brought up to enter the guide bush, and is advanced until the leading end of the piercing mandrel contacts the hot billet, which it slightly compresses before the actual piercing process starts. Immediately the container is completely filled with the compressed billet, the mandrel is forced under pressure of the main ram, through the billet, and with a small wad of metal—the core of the billet—enters the die. The main ram continues to advance until indication by calibration notifies the hydraulic valve operator that only a skull is left in the container. The holding wedge is then removed by a vertical ram mounted on the container buttress block, and the advance continued to bring the section under the control of the operators. The guide-pusher plate is released and returned to first working position, and the section is advanced further in order to sever the skull by means of cutting edges mounted on one of the horizontal rams.

The 8 ft. section extruded is drawn forward to a position clear of the actual working area and transferred to a cooling rack or bed from which it is taken to a reeler and straightening machine of the type shown in Fig. 2. During the time the section is being moved to the reeler, the various parts of the press are returned to position in readiness for the next billet

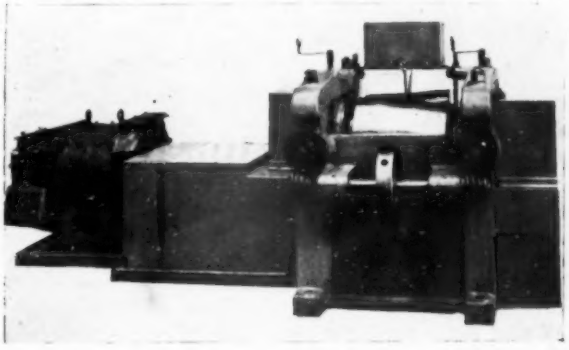
With the press under review, the actual time occupied in producing the shell section, from the time the billet is gripped by tongs on the furnace foreplate to the separation of the skull from the section, averaged 35 secs., the output per hour ranging from 33 to 36 shells, using the horizontal type of press indicated in Fig. 1.

The Extrusion Press

Apart from the pumps and accumulator, which are situated behind the guard wall shown at the rear of Fig. 1, it will be of interest to give a brief description of the horizontal extrusion press. Starting at the rear of the machine shown in the illustration is the cylinder and ram operating the piercing mandrel, and the main extruding cylinder and ram in the centre of two cylinders whose purpose is to extract main ram at the conclusion of the operation. A vertical cylinder integral with the end buttress is used for positioning and raising the wedge, which takes the piercing and full extrusion pressure. Two double cylinders, near to operators (Fig. 1) form

Fig. 2.—Direct driven reeling machine.

Courtesy of W. H. A. Robertson and Co., Ltd.



shears for the removal of the discard from the hollow section. The distance between the die-housing and the main ram crosshead, when retarded, is 9 ft. 9 in., the press having an overall length of 75 ft. The container which receives the billet is gas-heated. The casting forming the main cylinder weighs 19 tons, while the thrust buttress die-housing is 10½ tons. The total weight of pumps is 20 tons, while the total weight of accumulator cylinders is 29 tons. There are four main valves, and one valve for controlling the shear cylinders. The working pressure of the press is 3,000 lb. per sq. in.

With this plant solid billets are used, and it is not now considered necessary to skim them; both drilling and tooling of the billets have been eliminated, and it is claimed that, both in regard to quality and finish, a better tube is produced, while the concentricity gives a very close tolerance.

The vertical type of extrusion press has been installed by manufacturers of non-ferrous tubes especially for copper alloy tubes of finished diameter up to 3 in., in the United States and in Continental works. The writer has been privileged to demonstrate the satisfactory performance of this type of plant in several European countries, in some cases despite difficulties arising from the use of inferior materials.

The installation of the vertical press invariably causes discussions on the relative merits of the two types of presses and while it is not proposed to enter into any discussion on the subject here, it is general practice for manufacturers to keep to the floor level when space is available. With the vertical extrusion press the discharge end must be clear, and opinions differ on whether the press should be erected on a platform, say, 12 ft. high, above the floor level, or on the floor level, but over a pit about 12 ft. deep, into which the shells are discharged and from which they must be lifted to floor level. A number of other considerations are obviously involved in making a choice.

Comments on the Process with Horizontal Press

From the inception of the application of extrusion to the manufacture of tubes it has been the writer's contention that, where the area of the piercing tool allows it, pressure should be applied to the hot billet as soon as possible after its insertion in the container, in order that the 6¾ in. diameter billet will completely fill the 7 in. diameter container. With the guide-bush and plate a good sliding fit in the container and over the piercing and extrusion mandrel, the application of pressure may be expected to give satisfactory results.

An important factor, concerning the elimination of skimming and boring of the billets, is the space that would otherwise be involved for turning and drilling machines, and, incidentally, the scrap is in a form which facilitates handling. When in contact with some Continental plants before the war, this method was quickly put into operation following a demonstration by the writer. It saves initial outlay on expensive machinery, drilling of billets being carried out only when producing small hollow sections.

Die and Mandrel Material

A difficult problem in connection with extrusion is that of suitable steels for the die and mandrel. There are several steels manufactured for these purposes, but the difficulties encountered by the production engineer is frequently due to want of care in the use of the dies and mandrels. For the smaller diameter tubes especially it is important to make frequent changes, in particular the mandrel should be changed for each production operation. It must be remembered that, by comparison with the die, the mandrel is responsible for the piercing of the billet and also the compression of the heated material in the actual extruding process. In both die and mandrel, however, there is considerable friction at a time when these tools are heated by contact with the hot billet, and with the smaller shells the mass is relatively small.

In the modern presses discussed this problem has been overcome by the introduction of a mandrel holder which permits a quick release and enables changes to be effected between successive operations without appreciable loss of output.

The Reeling, Straightening and Releasing Machine

The writer is in no doubt that from the principle of this interesting process of reeling was evolved the first type of rotary piercing of round solid billets—the Mannesmann piercing mill. The excellent results achieved with the modern machine, however, are due to continued experimental work in connection with the shape of the contours of the two rolls, which are its main working components. The typical reeler shown in Fig. 2 illustrates the contours of these highly polished rolls.

The flexibility effected by horizontal adjustments of these main rolls makes it possible for the machine to deal with bars or tubes of varying diameters. The horizontal adjustment is effected through worm gearing, while the vertical adjustment of the guides which lie in the pass between the main rolls are used to raise or depress the tube or bar under operation and supported by a girder arrangement of simple construction. As in rotary methods, it will be noted that the axes of the main rolls intersect at an angle suitable for the material to be passed through. Although this machine is generally regarded as a means of straightening tubes or bars, the same process is used for tube drawing on the bar; it being used as a means of releasing the drawn section from contact with the mandrel. In the latter operation the pressure of the rolls slightly expand the tube, which permits the easy extraction of the mandrel bar. The improvements in the design and application of the reeling machine in recent years have kept pace with the developments in other plant used in tube manufacture.

Castings in Light Alloys

THE desirability of the closest possible co-operation between the founder and user, if the maximum advantage is to be taken of the very wide range of properties and potentialities of light alloy and non-ferrous castings, is emphasised in an excellently produced booklet issued by Kent Alloys, Ltd. The booklet contains an outline of facilities available at this company's works, of the scientific planning and control in the development, manufacture and inspection of light alloy castings. In addition, typical examples are given of sand castings and of pressure and gravity die-castings, while very useful charts are incorporated providing convenient references on compositions and properties of standard aluminium-base casting alloys and bronze casting alloys. Particulars are also given in connection with a number of aluminium alloy specifications.

Copies of this interesting and useful booklet may be obtained on application to Kent Alloys, Ltd. Commercial Road, Strood, Kent.

The Embrittlement of Ingot Iron and Low-Carbon Steels.

By B. Jones, D.Sc., F.I.C.

Abnormal structures occasionally found in low-carbon steels are difficult to interpret with the equilibrium diagram. In some cases they cannot be correlated with any particular temperature of heating, or with the rate of cooling, while the material was often associated with definite embrittlement. This feature, together with the brittle range of ingot iron, is discussed with reference to the equilibrium conditions at various temperatures and the effects of rapid rates of cooling from these temperatures.

THE physical and chemical properties of steels are generally determined by the composition and structure, but there are many examples of defective material, as judged by a lack of ductility, poor bend or low notched-bar impact properties, that cannot be correlated with these values. In high-carbon and alloy steels it has long been recognised that some steels of similar composition, as determined by standard methods of analysis, differ to a marked degree in their inherent properties, especially in their resistance to dynamic stresses and response to heat-treatment. Many vague terms have been used to describe this difference, among which the term " body " is the oldest and best known. It is now apparent that the chief factor in body is grain size, and another is ageing characteristics. The two are related in one respect, that both fine-grain and retarded ageing characteristics are produced by strong deoxidation of the steel, as with aluminium, for example. By grain-size control many of the beneficial properties induced by the additions of alloying elements can be obtained in plain carbon steels. At the same time, developments in engineering science have demanded steels with a high strength combined with appreciable toughness. As there is a narrow limit to the thickness of section that can be hardened satisfactorily by quenching carbon steels, alloy steels are necessary where lightness of weight is desirable, although, in certain cases these limitations may be overcome by the simple expedient of increasing the quantity of metal where weight is not of major importance, as in buildings and bridge work, etc.

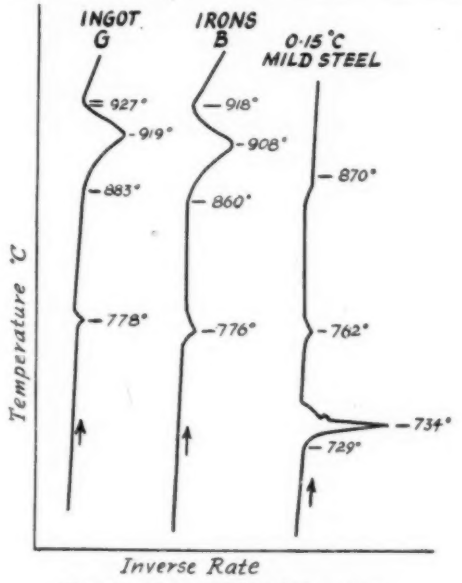


Fig. 1.—Thermal heating curves.

While low-carbon steels are seldom quenched and tempered, except for a few special applications, such as high-strength bolts and rivets, most alloy steels are generally employed in this heat-treated condition to obtain the full benefit of the additional elements. The largest output of steel is represented by plain carbon steels, mostly low-carbon, made by the basic open-hearth and Bessemer processes. Steels for strip, sheets, and forgings fall into this category. As with all steels, the heat-treatment processes are intimately bound up with the temperatures of the critical range—namely, the temperatures of the Ac_1 and Ac_3 critical points. Much of the output of low-carbon steel is of rimming steel, in which the surface is of lower carbon content and, at least in the ingot, of greater density and freedom from sponginess than the centre. Strength is seldom a matter of concern in these steels, the prime aim being ability for severe deformation. Hence the heat-treatment given is for grain refinement and maximum ductility. This is effected by normalising or annealing, the latter treatment being carried out at a temperature above the Ac_3 or at

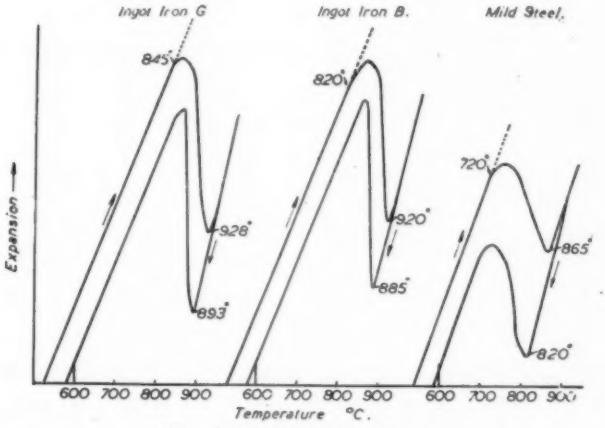
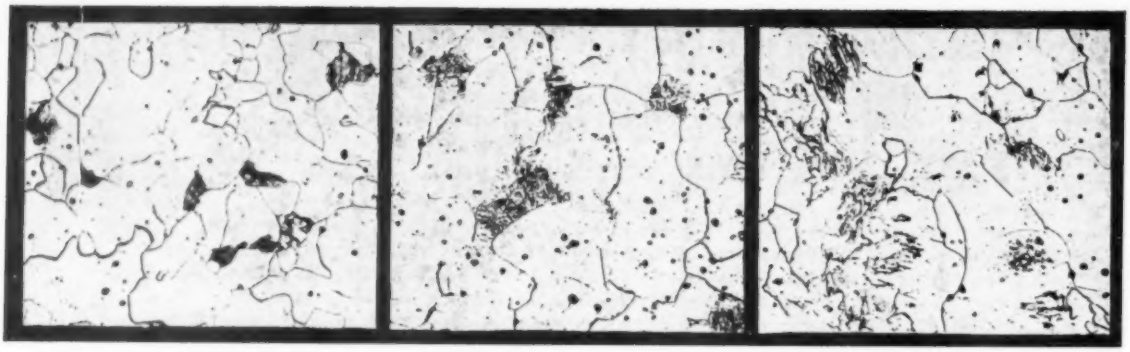


Fig. 2.—Dilatometer curves.



(A) B. W.Q. 850° C. x 100. (B) G. W.Q. 875° C. x 100. (C) G. W.Q. 900° C. x 100.

(D) B. W.Q. 850° C. x 350. (E) G. W.Q. 850° C. x 500. (F) G. W.Q. 875° C. x 350.

(G) B. W.Q. 875° C. x 350. (H) G. W.Q. 900° C. x 350.

Fig. 3.—Photomicrographs of rapidly cooled ingot irons (½-in. sections).

a sub-critical heat for severely cold-worked material, to obtain recrystallisation and spheroidising of the carbide.

The Structures of Low-carbon Steels

Many instances are known in which abnormal structures have been found in low-carbon steels which have often been difficult to interpret with the equilibrium diagram. This is particularly the case in the non-pearlitic steels, containing less than 0·04% carbon. This class includes ingot iron and rimming steels for cold-pressing operations. There has been a marked tendency in recent years to reduce the carbon content of these steels to low values to obtain the most beneficial results for cold pressing. Many of the structures found could not be correlated with any particular temperature of heating, or with the rate of cooling, while the material was often associated with definite embrittlement. This was shown by abnormally low bend test results, or with low impact values, as shown in a paper by Jones and Gray.[1] The well-known brittle range of ingot iron when hot-rolled in a range of approximately 900°–800° C. a s been the subject of much controversy for many years, and there is still no agreement as to the cause.

Embrittlement of mild steel is often due to ageing effects, which cause an increase in hardness at atmospheric temperatures with time, the process being accelerated at slightly higher temperatures. Strain frequently plays such a prominent part that it is helpful to differentiate between ageing which results from heat-treatment and that which occurs after cold work. It is proposed to discuss these various features, particularly with reference to the equilibrium conditions at various temperatures, and the effects of rapid rates of cooling from these temperatures.

The Critical Range of Low-carbon Steels

The constitution of the non-pearlitic steels has been incompletely understood until recent years, as shown by Whiteley[2] and by Jones and Gray.[1] This is due to the

1 Jones and Gray. "Martensitic Grains in Rapidly Cooled Ingot Iron and Mild Steel." *Journal of the Iron and Steel Institute*, 1938, vol. I, p. 327P.
2 Lea and Arnold. "The Embrittlement of Low-carbon Steel." *Proceedings of the Institution of Mechanical Engineers*, 1934, vol. 127, p. 299.

fact that the early diagrams of the iron-carbon system were constructed by joining up the temperatures of the critical points which were found by thermal analysis. For this reason, the solid solubility line of carbon in ferrite, which is a feature of recent diagrams, was not found for many years, until Koster[4] published the modified diagram for mild steels in 1929. In recent years the dilatometer has been used to determine the temperatures of the critical range with more precision than by thermal analysis, as it indicates the commencement and completion of the $\alpha - \gamma$ iron transformation more sharply. This is particularly the case in the pearlitic mild steels, in which the end of the critical range, or Ac_3 critical point, is very ill-defined in an inverse-rate heating curve. This is shown in Fig. 1, in which the Ac_3 point of a 0·15% carbon steel is indicated only by a slight change of direction in the heating curve. In the absence of special apparatus, which gives a uniform heating rate, the point may not be shown in the curve. The dilatometer curve (Fig. 2) shows a sharp end point. The other curves in Figs. 1 and 2 refer to ingot irons; ingot iron G = C 0·020%, Mn 0·05%, Cu 0·07%; ingot iron B = C 0·020%, Mn 0·05%, Cu 0·30%, Ti 0·01%; steel W = C 0·15%, Mn 0·56%.

In low-carbon iron and steel it is not always appreciated that the Ac_3 critical point extends over a range of temperature, which is often of large dimensions and is of special significance. The temperature of the critical point is usually given as the maximum, or mean of this temperature range. In the author's view, it is more correct to take the end of the so-called critical point to indicate the Ac_3 temperature, which agrees with the results obtained by the dilatometer method. Actually, the thermal arrest at the so-called "Ac_3 critical point" in non-pearlitic steels indicates allotropic changes that are more complex in character than are generally realised, and they are of the utmost importance. Ignoring the Ac_2 magnetic change-point, at approximately 770° C., which has no effect on heat-treatment technique, the Ac_3 temperature range found in continuous heating curves indicates, to some extent, the commencement of the formation of γ-iron or austenite, and its completion. Although the dilatometer curves show the transformation of ferrite to austenite more sharply, and at lower temperatures than by thermal analysis, the lower critical temperature in non-pearlitic steels is not *accurately* revealed in continuous heating curves by either method, even with slow rates of heating.

This is shown by quenching tests on small specimens after soaking at various temperatures, which reveal the first presence of austenite quite readily. The $\alpha - \gamma$ change in non-pearlitic steels is equivalent to the Ac_1 point of pearlitic steels, which is also readily revealed by quenching tests from 730° C. Both changes are shown by the presence of martensite in the microstructure. The author has proposed the nomenclature of ac_1 to indicate the lower critical temperature in the non-pearlitic steels, to distinguish it from the Ac_1 temperature of pearlitic steels. The temperatures of the ac_1 transformation for low-carbon materials are shown by the line DP in the equilibrium diagram given in Fig. 5, while the end of the critical range, or Ac_3 temperature, is shown in the line GO. As the purity of iron is increased, the temperature range between ac_1 and Ac_3 is reduced, but as pure iron free from carbon has not yet been made, it is not confirmed whether the two temperatures would ultimately coincide at zero carbon. For this reason, the points G and D are retained, and are separated in the equilibrium diagram at a slight carbon content.

Effects of Rapid Cooling

The solubility area of carbon in ferrite, denoted by DPA in Fig. 5, has great significance in the treatment of low-carbon steels. The reduced solubility of carbon at low temperatures causes dispersion hardening in steels that are cooled rapidly from elevated temperatures. This age-hardening effect is referred to later. Quenching of mild steels from above Ac_3 is sometimes carried out on fairly thick sections to raise the tensile strength, without loss of ductility and toughness as found with steels with higher carbon contents. Even very low-carbon steels will harden to some extent by this treatment. For special applications, such as for the production of high-strength bolts and wire rod for upholstery springs, low-carbon steels are water-quenched from above the Ac_3 temperature. Mild steels have high critical cooling rates, and therefore need drastic quenching in practice for maximum results. This is obtained by cooling them in rapidly moving water, or by means of a water spray in special equipment. The results of quenching two ingot irons and a 0·15% carbon steel are shown in Tables I–III. The quenching tests were carried out on bars $\frac{1}{2}$ in and $1\frac{1}{2}$ in thick.

(*a*) *Ingot Iron.*—The mechanical properties of ingot irons, apart from the limit of proportionality, are remarkably constant and are practically unaffected by either the mass or the temperature of quenching from 650°–1,000°C. The hardness increases on quenching them from temperatures up to 760° C., but it is then constant, or actually falls until 900° C. is reached. The remarkable

3 Whiteley. *Journal of the Iron and Steel Institute*, 1925, vol. 1, p. 315.
4 Köster. *Archiv. für das Eisenhüttenwesen*, 1929, vol. 2, p. 503.

TABLE I.—QUENCHING OF INGOT IRON G (C 0·020%, MN 0·05%, CU 0·07%).
MECHANICAL PROPERTIES.

Quenching Temperature, °C.	Limit of Proportionality, Tons per sq. in.		Yield Stress,† Tons per sq. in.		Maximum Stress, Tons per sq. in.		Elongation on 3 in. %.		Reduction of Area, %.		Impact Value, Ft.-lb.		Hardness.	
	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)
As normalised	8·2	—	9·0	—	21·2	—	42·0	—	70·8	—	137	—	93	—
650	9·0	7·2	10·5	10·7	22·2	22·6	43·0	43·0	70·5	69·9	140	132	110	115
700	9·7	5·7	10·8	11·1	22·7	22·7	41·5	40·5	67·5	68·9	144	136	113	112
730	9·5	5·7	11·0	11·0	22·7	23·5	41·2	41·0	67·8	69·0	124	123	117	120
760	9·6	6·5	11·0	11·1	23·0	22·8	40·7	41·0	67·1	67·0	125	121	120	119
800	9·4	5·4	11·5	11·0	23·0	22·6	41·5	41·5	66·2	69·2	142	134	111	118
825	9·4	6·4	11·2	11·2	22·8	22·8	41·0	39·8	65·0	66·9	134	136	110	121
850	9·5	7·4	10·9	11·7	22·5	23·0	40·7	39·7	63·0	64·0	126	135	109	118
875	—	—	—	—	—	—	—	—	—	—	71	11	108	117
900	10·2	6·2	11·4	12·3	22·9	23·7	39·5	39·5	61·3	67·6	57	6	120	130
925	—	—	—	—	—	—	—	—	—	—	94	105	120	130
950	9·3	7·1	11·5	11·8	22·7	22·8	37·5	36·5	59·8	65·8	124	115	122	132
1,000	9·7	5·6	12·0	12·0	22·9	22·9	38·0	36·0	57·5	62·7	95	90	125	131

(1) = $1\frac{1}{2}$-in. Bars. (2) = $\frac{1}{2}$-in. Bars. † Load giving 0·2% permanent set.

(A) W.Q. 730° C. x 350. (B) W.Q. 730° C. x 1,000. (C) W.Q. 730° C. and T. 600° C. x 350.

(D) W.Q. 760° C. x 350. (E) W. Q. 770° C. x 500. (F) W.Q. 800° C. x 350.

(G) W.Q. 800° C. x 500. (H) W.Q. 850° C. x 350. (I) W.Q. 900° C. x 350.

(J) W.Q. 1,000° C. x 350. (K) W.Q. 730° C. (1¼-in. bar) x 350. (L) W.Q. 770° C. (1¼-in. bar) x 350.

Fig. 4.– Photomicrographs of rapidly cooling 0.15% C. steel (½-in. sections, except K and L).

TABLE II.—QUENCHING OF INGOT IRON B (C 0·020%, MN 0·05%, CU 0·30%, TI 0·01%).
MECHANICAL PROPERTIES.

Quenching Temperature, °C.	Limit of Proportionality, Tons per sq. in.		Yield Stress,† Tons per sq. in.		Maximum Stress, Tons per sq. in.		Elongation on 3 in. %		Reduction of Area, %		Impact Value, Ft.-lb.		Hardness.	
	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)
As normalised	12·8	—	14·0	—	21·3	—	43·5	—	75·0	—	185	—	98	—
650	9·6	5·6	12·7	11·8	23·0	23·6	39·5	39·0	68·4	69·0	146	140	119	129
700	12·8	8·7	13·5	14·5	23·4	25·2	39·5	34·0	67·6	65·6	146	130	124	127
750	12·3	8·5	13·8	15·4	24·0	25·3	39·0	38·0	67·0	66·2	137	135	127	131
760	11·5	7·0	13·0	13·0	24·0	24·5	42·0	38·0	69·6	67·1	132	145	130	132
800	12·0	7·5	13·4	13·3	24·0	24·2	40·0	40·0	67·9	67·9	174	140	117	128
825	11·8	7·7	13·2	13·7	24·0	24·4	39·0	39·0	66·5	66·0	166	150	118	133
850	11·6	8·0	12·8	14·0	24·0	24·5	38·0	38·5	65·6	66·0	158	143	113	128
875	—	—	—	—	—	—	—	—	—	—	75	10	113	128
900	13·7	6·7	13·8	13·8	24·5	25·0	34·5	33·0	65·0	62·5	103	45	128	139
925	—	—	—	—	—	—	—	—	—	—	125	123	125	140
950	13·6	6·0	14·8	14·0	23·9	25·3	37·0	37·0	64·7	61·5	172	160	127	142
1,000	14·0	5·0	14·9	14·8	24·2	25·6	34·3	33·0	62·4	62·0	160	129	129	147

(1) = 1¼in. Bars. (2) = ½-in. Bars. † Load giving 0·2% permanent set.

feature is shown by the extremely low-notched bar impact values after quenching from temperatures of 875° and 900° C., particularly in the small sections. The materials are tough when quenched from 850° and 950° C., while intermediate values are obtained from 925° C. There appears to be no correlation with any other property, as the ductility of the iron is normal. The microstructures show evidence of martensite on quenching

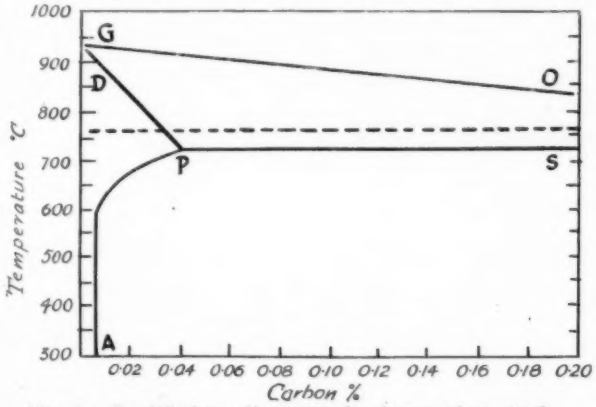


Fig. 5.—Equilibrium diagrams for low-carbon steels.

from temperatures above the ac_1. Typical structures are shown in Fig. 3. Austenite is formed in the ingot irons above the ac_1 temperature, which is at 830° C. in G, and at 800° C. in the copper-titanium iron B. The embrittlement is not due to the presence of the relatively high-carbon martensite areas shown in Figs. 3A, 3D, and 3E, however, but to the larger areas of low-carbon martensite, formed from the austenite that had increased in volume by spreading along the grain boundaries as the Ac_3 temperature was approached. Thus, on heating ingot iron with 0·02% carbon, the iron carbide is taken into solution at a temperature where the composition line cuts the line AP, shown in Fig. 1, and the austenite areas formed at 830° C. increase in volume continuously up to 928° C. by the absorption of ferrite. The microstructures, after quenching from 875° C. and 900° C., are shown in Figs. 3B, 3C, 3F, 3G, and 3H. These show large martensitic areas, which are formed from austenite, containing less than 0·12% carbon, and cause embrittlement after quenching. The structures obtained on quenching from 900° C. (875° C. in the case of ingot iron B) are characterised by parallel wavy lines that cause a splitting up or "buckling" of the grain boundaries. The low-carbon austenite at 900° C. thus forms a dissociated type of martensite that causes an abnormal condition of the grain boundaries. It will be observed that the temperature of the brittle range between 850° and 950° C., found by quenching, also coincides with the brittle range found on hot-rolling the materials. The same cause may apply to both processes—namely, the disturbance of the grain boundaries by the rapid diffusion of austenite through the mass of metal over a narrow range of temperature which may be accentuated by the presence of iron sulphide.

(*b*) *Mild Steel.*—Table III shows that there is a brittle range of temperature on quenching, between 700° and 800° C., which results in very low impact values, particularly from 730°–750° C. This was also found by Lea and Arnold.[2] The effect of the thickness of section is obvious, as the slower rate of cooling of the 1-in. bars results in higher impact values. The toughening of these bars on quenching from 850°–900° C. is well marked, the impact results being higher than those obtained by normalising, while, in addition, high values of yield stress and tensile strength are obtained. The brittle and tough conditions are intimately bound up with the temperatures of the critical range shown in the

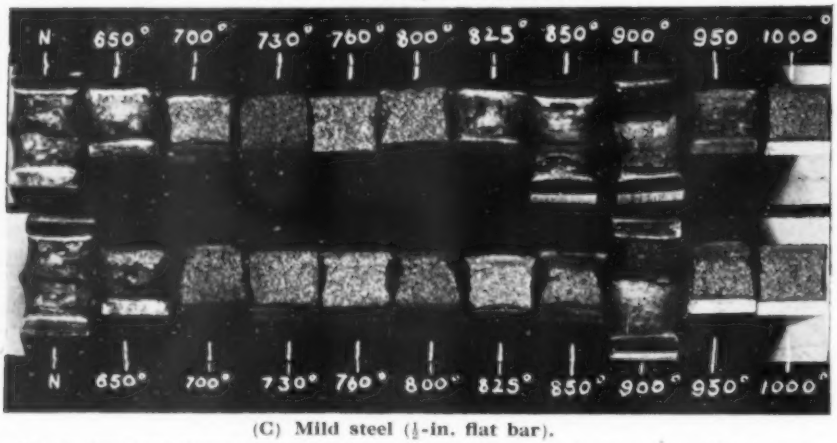
TABLE III.—QUENCHING OF MILD STEEL (C 0·15%, SI 0·05%, MN 0·56%, CU 0·07%).
MECHANICAL PROPERTIES.

Quenching Temperature, °C.	Limit Proportionality, Tons per sq. in.		Yield Stress,† Tons per sq. in.		Maximum Stress, Tons per sq. in.		Elongation on 3 in. %		Reduction of Area, %		Impact Value, Ft.-lb.		Hardness.	
	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)	(1)	(2)
As normalised	17·5	—	18·0	—	27·6	—	38·0	—	65·4	—	125	—	131	—
650	15·1	9·2	17·7	16·6	28·8	29·6	37·5	36·0	65·3	64·9	89	82	142	148
700	17·0	9·6	17·9	18·8	28·8	31·1	36·5	33·7	65·3	64·9	51	25	145	160
730	13·8	9·4	17·8	22·3	34·9	38·9	30·0	24·7	61·0	48·0	23	8	161	207
760	13·0	7·6	18·1	21·4	35·0	40·6	30·0	21·0	60·5	41·0	12	10	179	216
800	13·7	8·3	18·2	22·0	32·4	41·5	32·0	15·0	66·3	42·0	18	16	172	207
825	12·4	7·2	18·0	20·4	32·0	37·4	31·0	19·0	66·5	56·5	93	28	161	198
850	15·6	9·2	21·6	23·2	33·5	38·9	35·5	25·2	68·5	65·7	138	38	167	185
900	15·4	8·4	22·4	25·5	34·6	41·5	27·0	20·0	66·0	51·3	132	46	171	210
950	14·2	8·4	23·0	31·0	35·0	48·6	25·5	14·8	61·5	34·0	68	21	190	232
1,000	12·1	7·6	23·0	33·8	37·3	52·5	24·5	9·0	61·0	20·0	26	13	201	286

(1) = 1¼-in. Bars. (2) = ½-in. Bars. † Load giving 0·2% permanent set.

(A) Ingot iron G ($\frac{1}{2}$-in. flat bar). (D) Mild steel ($\frac{1}{2}$-in. flat bar).

(B) Mild steel ($1\frac{1}{8}$-in. round bar).



(C) Mild steel ($\frac{1}{2}$-in. flat bar).

Fig. 6. Impact fractures of ingot iron and 0.15% C steel after normalising and water quenching at temperatures given. (N = normalised, T = tempered.)

equilibrium diagram. The brittle condition is found on quenching from temperatures close to the Ac_1 point, and coincides with the presence of martensitic grains at the grain boundaries of the ferrite. This is illustrated in Figs. 4A and 4B. On heating to 730° C. the austenite formed is converted into almost structureless martensite, which is surrounded by a border of troostite. Quenching of sections smaller than $\frac{1}{4}$ in. results in the production of martensite without the troostite band on quenching from 730° C., while the quenching of the thicker $1\frac{1}{8}$ in. bars gives troostite areas only in the ferrite, as shown in Fig. 4K. The martensitic or troostitic areas form an almost continuous network around the ferrite grains, and this structure corresponds to the maximum embrittlement. Structures of the steel after quenching from higher temperatures are also shown in Fig. 4. Above the Ac_1 temperature, the austenite absorbs the adjacent ferrite grains and becomes lower in carbon content; on quenching, it forms a brown-etching type of dilute martensite. Quenching from 800° and 825° C. imparts a "dog-tooth" pattern to the martensite, characterised by sharp angular boundaries (Fig. 4F and 4G), which is also associated with low impact values. Heating to 850° C. results in the absorption of most of the ferrite as the steel approaches the end of the critical range, or Ac_3 temperature, and, on quenching, a more homogeneous structure of dilute martensite is formed (Fig. 4H). At 900° C. the formation of austenite is complete through the metal, and, on quenching, the dark-etching structure of Fig. 4I is obtained. Overheating the steel to 950°–1,000° C. produces a coarse acicular martensite, which again causes low impact values.

The effect of tempering mild steel, after quenching from 730° C., is shown in Table IV. The steel is toughened even at 200° C., while above 400° C. the troosto-martensite areas are converted to sorbite, as shown in Fig. 4C.

The impact fractures obtained after quenching ingot iron and mild steel from various temperatures are shown in Fig. 6. They show the production of "crystalline" brittle fractures after quenching from the critical range of temperatures, and the recovery of toughness on quenching from higher temperatures.

TABLE IV.

TOUGHENING OF EMBRITTLED MILD STEEL QUENCHED FROM 730° C., BY TEMPERING

Heat-treatment.	Impact Value Ft.-lbs	Hardness.
As normalised	125	131
W.Q. 730° C.	8	203
W.Q. 730° C., T. 100° C.	6	179
,, T. 200° C.	27	162
,, T. 400° C.	97	157
,, T. 600° C.	141	127

The results of the tests have explained many failures found in low-carbon steels. The dangerous range of temperature has special significance with regard to the toughness of the core metal in case-hardening steels which are quenched from a single temperature after carburising. Embrittlement found by water-quenching comparatively thick sections may also be revealed by the air cooling of thin sections, such as steel sheets, from similar temperatures. J. A. Jones has shown, in a discussion of the work of Jones and Gray,[1] that the air-cooling of a mild steel containing 1·5% manganese from a temperature of 700°–750° C. causes a marked embrittlement similar in character to the water-quenching of low-manganese mild steel. Conditions leading to embrittlement in engineering practice have been found in riveting, and in work suddenly cooled by quenching in a bosh of water. Cases have been known of water-cooling low-carbon steel and wrought-iron forgings during shaping processes, with consequent strange failures in service.

Ageing of Mild Steel

The properties of low-carbon steels are considerably influenced by ageing at atmospheric temperatures, or at temperatures up to 250° C., when in a state of stress which has been caused either by quenching or cold-work. The processes are known as quench-ageing and strain-ageing, and when embrittlement occurs, the conditions

5 Andrew and Trent, *Journal of the Iron and Steel Institute*, 1938, vol. 11, p. 241P.
6 Stromeyer, *ibid*, 1907, p. 200.

are known as quench-age embrittlement and strain-age embrittlement respectively. Stromeyer,[6] in 1907, noted embrittlement in course of time in basic Bessemer boiler plate, which he attributed to the presence of nitrogen. Ageing has been identified as the cause of the " blue-heat phenomenon " in iron and steel, which refers to the increase in strength at about 200° C., which was made more marked by the presence of nitrogen. Recent interest has been increased by the observation that ageing effects can be minimised by drastic deoxidation, which suggests a relationship between grain size and ageing. In so far as embrittlement in the notched bar impact test is concerned the relationship is probably direct, since grain size affects the impact properties. Otherwise, the relationship is probably indirect.

(*a*) *Quench-ageing.*—Ageing which results from heat-treatment is related in general to strain-ageing, in that a steel which shows quench-ageing will also show strain-ageing. Koster,[4] in studying the effects of time and temperature on the ageing of mild steel after quenching from 680° C., found that the hardness increased more rapidly by ageing above room temperature, but that the maximum hardness attained was lower. At the higher ageing temperatures the hardness began to fall after a short time, but at room temperature the hardness kept on increasing. The degree of quench ageing is greatly reduced by strongly deoxidising the steel, particularly with aluminium. Swinden and Bolsover[7] showed that there was no appreciable fall in the impact test in a fine-grained mild steel after quenching from 680° C. and ageing, whereas in a coarse-grained steel there was a drop in impact value of about 50%. Andrew and Trent[5] confirmed the results of previous work that the maximum age-hardening occurred in a steel with a carbon content of 0·03–0·04% after quenching it from a temperature just below the Ac_1 temperature. They also showed that nitrogenised iron, after quenching, gave a greater increase in hardness than iron-carbon alloys, and that nitro-martensite, unlike martensite in an iron-carbon alloy, showed considerable ageing. As in the ageing of duralumin, the exact stage of precipitation at which hardening takes place is not completely understood, but the view is held that the age-hardening is due to the strain in the lattice produced by carbide or nitride of iron, prior to their precipitation from solid solution.[5] Nitrogen is much more soluble in iron than carbon, as shown in Fig. 7, which shows the iron-nitrogen diagram superimposed on the iron-carbon diagram. On quenching iron with a high nitrogen content, a greater hardness is attained on ageing; thus a Bessemer rimming steel will tend to age-harden to a greater extent than open-hearth steel, particularly if segregation of nitrogen occurs.

An application of the quench-ageing of mild steel is in the production of upholstery spring wire. The method was first applied about 50 years ago by producing a wire generally known to the trade as " water-hardened." The steel, containing less than 0·1% carbon, is quenched in water from a temperature above the Ac_3 point and aged. It is then drawn into wire. The qualities demanded are a high tensile strength and the ability to withstand knotting of the wire in the cold after the last draw.

(*b*) *Strain-ageing.*—In practice, the effects of strain-ageing are probably of greater importance than those of quench-ageing. The jog in the stress-strain curve at the yield-point in a tensile test manifests itself in the drawing and stamping of mild steel sheets as " strecher strains," giving an undesirable roughening of the surface. The jog is a characteristic of iron and steel in the soft-annealed condition, and can be eliminated by slight deformation as in the final " skin rolling " given annealed sheet for this purpose and for straightening. On ageing, the jog may return with time so that after storage the sheet may regain the tendency to strencher strain. This does not occur in drastically deoxidised non-ageing sheet. Extreme brittleness has been found in cases of formed mild steel sheets which were subsequently galvanised, during which process the slightest pressure caused the steel to fracture in a brittle manner. Pieces subsequently cut from the article were bent over and hammered flat with no sign of rupture. Tests show that if dead-soft steel is stretched more than 5% of its length and subsequently reheated to a temperature of 200°–300° C. the brittleness could be reproduced.

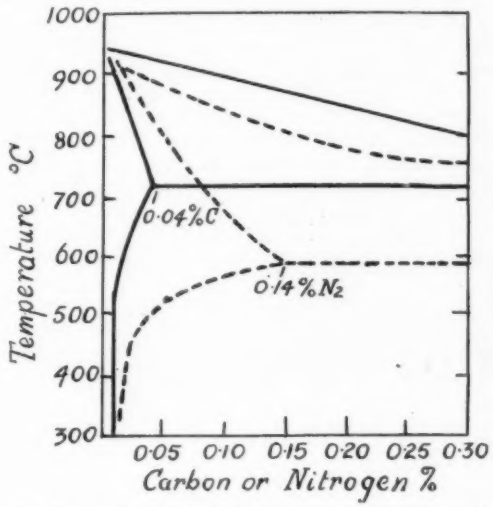


Fig. 7.—Iron-nitrogen diagram superimposed on iron-carbon diagram.

Some precipitation may occur during the straining itself, and it is probably for this reason that cold-working alone produces a greater increase in hardness in " ageing " than in " non-ageing " steels. The property of mild steel of developing embrittlement, as judged by the impact value, following the reheating of the steel to a temperature of 250° C. after straining 15% was investigated by Swinden and Bolsover.[7] It was shown that the resistance to strain-age embrittlement was considerably increased in a fine-grain steel.

Edwards and his colleagues[8] have proved that elements that show definite tendencies to form carbides lessen the extent of strain-ageing, while those elements that show little or no tendency to combine with carbon increase the degree of strain-ageing. Andrew and Lee[9] have recently shown that the properties of a cold-worked steel are similar in many respects to those of a steel of the same composition in the quench-hardened condition. They have given evidence to show that carbon is dissolved to form austenite during the work-hardening process, and that under certain conditions both austenite and martensite are formed.

7 Swinden and Bolsover. *ibid*, 1936, vol. 11, p. 457P.
8 Edwards, Phillips and Jones. *ibid*, 1940, vol. 11, p. 199P.
9 Andrew and Lee. *ibid*, 1942, vol. 1, p. 153P.

100,000-Hour Creep Test

It is generally recognised that the results of creep tests depend upon the method of testing employed as well as on the conditions of the test with regard to loading, temperature and time. Opinions differ on the application of test results to the design of parts for service and high temperatures, especially results from short-time tests. In this connection it is noteworthy that E. L. Robinson reports the results of a 100,000-hour test, data from which is reproduced here.*

A SERIES of 100,000-hour creep tests has been completed by the General Electric Company, N.Y., on four bars of S.A.E. 4330 type nickel-chrome-molybdenum steel. The bars had been continuously at a temperature of 450° C. throughout the test period, except for a brief period in 1935, when, after 37,000 hours, the furnace was shut down and moved to another building.

Elastic extensions are shown in Fig. 1. These lines serve to determine the elastic modulus for each of the several bars. These values are well defined, since, because of the removal of the furnace, each bar was loaded and unloaded twice. It did not seem possible to represent the elastic behaviour of all four bars by a single value of the modulus.

The curves of increasing plastic extension throughout the entire interval are shown in Fig. 2, while the chemical composition, heat-treatment and physical properties of the steel at room temperature are given in Table I.

TABLE I.—S.A.E. TYPE 4330 STEEL.

CHEMICAL COMPOSITION, %.

C.	Ni.	Cr.	Mo.	Mn.	Si.
0·31	2·05	0·83	0·45	0·54	0·11

HEAT-TREATMENT.

825° C. furnace cool, 675° C. oil quench, 425° C. furnace cool.

PHYSICAL PROPERTIES.

Tensile Strength, Lb./sq. in.	Elastic Limit, Lb./sq. in.	Elongation, %	Reduction of Area, %
111,800	83,000	22·5	59·5

ELASTIC MODULUS AT 450° C.

Specimen 281 : 23,500,000 lb./sq. in.
" 282 : 26,600,000 "
" 283 : 25,900,000 "
" 284 : 27,600,000 "

* *Mechanical Engineering*, March, 1943, pp. 166–8.

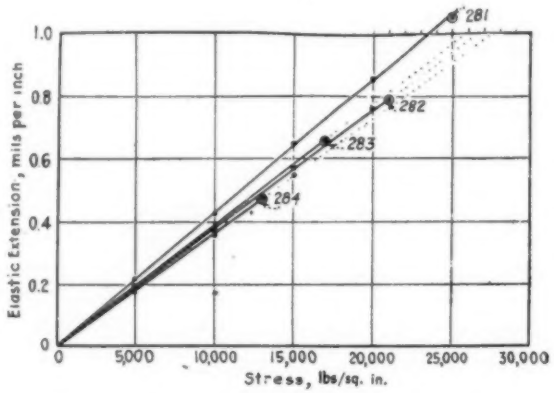
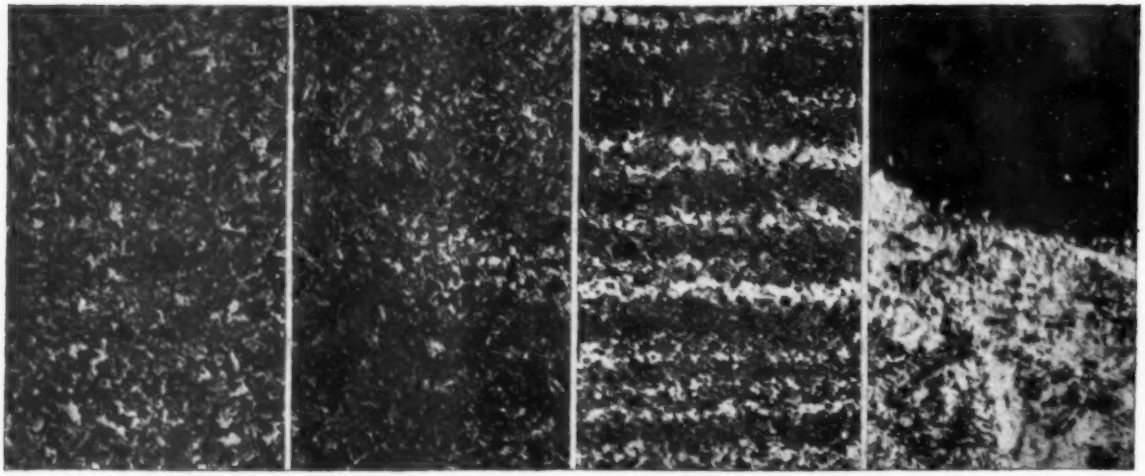


Fig. 1.—Elastic extensions immediately recoverable on removal of load.

(Because the furnace was shut down and moved at 37,000 hours each bar was loaded and unloaded twice, giving four separate measurements of the modulus of elasticity for each bar.)

Fig. 3 shows photomicrographs of the material, A in its original condition, and B and C for the two most highly stressed bars after 100,000-hour period at high temperature. All these bars came from the same heat of steel. D shows the oxide film at the conclusion of the test. The ends of this bar which had not been subjected to high temperature had a diameter of 0·5049 in. The part of the bar in the furnace measured 0·5075 in. diameter



A.—Original condition. x 100.

B.—Specimen 281 after 100,000 hours at 450° C. under 25,000 lb.±sq. in. x 100.

C.—Specimen 282 after 100,000 hours at 450° C. under 21,000 lb.±sq. in. x 100.

D.—Oxide film on surface of specimen 282 after 100,000 hours at 450° C. under 21,000 lb.±sq. in. x 100.

Fig. 3.—Photomicrographs of the material.

over the oxide film and 0·5032 in. with the film recovered down to bright metal. Surface penetration thus amounted to less than 1 mil. in 100,000 hours.

Fig. 4 shows several relations between stress and creep rate. It will be noted that the bar under 25,000 lb./sq. in. finished the 100,000-hour test with practically the same creep rate it had at 2,000 hours. However, during a number of the intervening years it had elongated at a rate but little more than half as fast. On the other hand, the bar under a load of 17,000 lb./sq. in. finished the test with its minimum rate, and the graph would suggest that the bar under 13,000 lb./sq. in. had not yet attained its minimum creep rate.

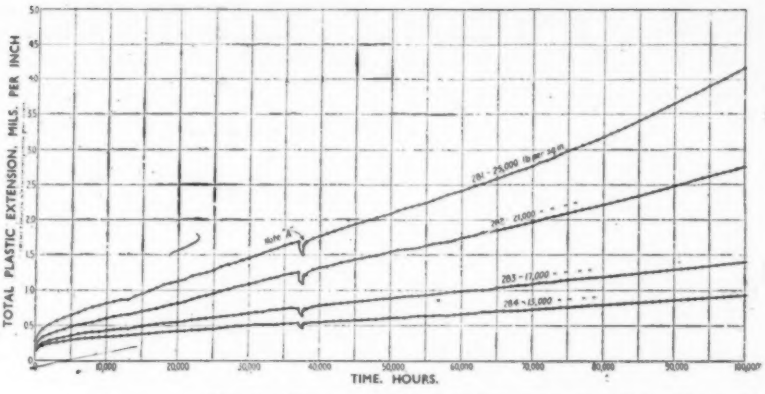


Fig. 2.—Curves of total plastic extension throughout the 100,000 hours at 450° C. Elastic extension is not included, (see Fig. 1.)

(Note A: After 37,000 hours the loads were dropped to 1,000 lb./sq. in. for a period of 570 hours after which the furnace was shut down, moved to a different building, and started once more. These curves show plastic recovery only.)

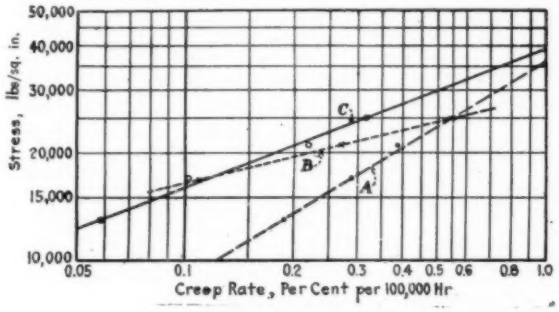


Fig. 4.—Creep rates at 450° C.

Line A.—After first 2,000 hours.

Line B.—Final rate after 100,000 hours.

Line C.—Minimum rate during entire 100,000 hours. This line also represents the rate with a total elastic-plus-plastic extension of 2 mils. per in.

Since these are all constant-stress tests, an estimate of probable performance under relaxation conditions with an elastic-plus-plastic extension of 2 mils. per in. may be made by noting the creep rate on each curve in Fig. 2 at a plastic extension less than 2 mils. per in. by the amount of the elastic extension shown in Fig. 1. This procedure leads to a point on each curve in the region of minimum rates and thus gives a line in Fig. 4 coinciding with line C for minimum rates.

The first four years of these tests were described in a paper by S. H. Weaver, entitled "The Creep Curve and Stability of Steels at Constant Stress and Temperature," Trans. A.S.M.E., vol. 58, November, 1936, pp. 745–751.

X-Ray Analysis in Industry

SOME important developments in the methods and technique of examining the behaviour of solids by X-ray diffraction methods were announced at the Institute of Physics' second conference on X-ray Analysis in Industry, which was held in Cambridge on April 9 and 10 last. The first session was devoted to a discussion on "The Quantitative Treatment of Powder Photographs," the introductory papers being "The Cylindrical Specimen," by Dr. A. J. Bradley, "The Flat Specimen," by Dr. G. W. Brindley, and "Microphotometers," by Professor J. M. Robertson. The second session opened with a brief discussion on a proposal to change the present values used for X-ray wave-lengths to absolute values. This was followed by a discussion on "Some Recent Developments in X-ray Diffraction." Dr. A. H. Jay spoke on "Superlattice Formation," Dr. K. Lonsdale on "Thermal Effects and Allied Phenomena," and Dr. H. Lipson on "Side-band Formation." Dr. A. J. C. Wilson closed the session by giving an interim report on the very satisfactory progress that had been made in the preparation of the Index to X-ray Diffraction Photographs of Powders, the preparation of which had been proposed at the first conference. The last session was devoted to a discussion on "Line Broadening." Dr. F. W. Jones opened it with a paper on "Quantitative Treatment of Crystal-size Measurement," and was followed by Dr. A. J. C. Wilson, who spoke on "The Effects of Distortion and Other Imperfections."

Mr. C. W. Bunn, of Imperial Chemical Industries Ltd., gave a most stimulating evening address with the title, "Towards Atomic Photography," which he illustrated with many beautiful slides. He spoke of some of the work which has been proceeding in the Cavendish Laboratory and of his own contributions to the development of new techniques.

The conference, over which Sir Lawrence Bragg presided, was attended by over 200 persons. The physicists, chemists, metallurgists and engineers who were present came from many different industries and Government establishments, as well as from academic institutions, thus demonstrating the growing field of application of this method of analysis. The organising committee of the conference was requested to seek the sanction of the Board of the Institute of Physics for the formation of a new group under its auspices so that the work of these first two conferences might be continued.

Heat-Treatment of the Wrought Aluminium Alloys

Success in obtaining the maximum mechanical properties from heat-treatable aluminium alloys depends largely upon a strict attention being given to details throughout the various operations. In this outline of heat-treatment practice some important aspects are stressed.

THE wrought aluminium alloys are divided into two main groups according to the means by which their maximum properties are obtained—namely, by work-hardening only or by a combination of working and heat-treatment. Work-hardened alloys are strengthened only by cold work, such as rolling or drawing, and any alloy can be obtained in the "soft" condition or in varous "tempers" up to hard, intermediate strengths being indicated by half-hard, quarter-hard, etc. The group of alloys strengthened by heat-treatment can be sub-divided into two classes, those which are solution-treated and those precipitation-treated or aged.

Full heat-treatment comprises the heating of the alloy to a specified temperature, cooling it rapidly from this temperature, and ageing, either spontaneously at ordinary temperatures or as a result of low temperature re-heating. The first two operations embrace solution-treatment, while the spontaneous hardening of certain alloys is known as age-hardening to distinguish it from the low temperature operation, which is referred to as precipitation treatment.

Solution Heat-Treatment

The maximum hardening effect is obtained only when the solution-treatment temperature is as high as practicable. But the upper temperature permitted is limited by the risk of partial melting of the metal. Unduly long soaking periods are required when the temperature is too low, and even then low mechanical properties are obtained. The temperature used in practice must be below the theoretcal maximum, due to slight variations in the composition of the alloy, and to lack of uniformity in the temperature distribution within the furnace. Figures specified for a particular alloy must therefore be rigorously followed.

Severe over-heating is revealed as blisters on the surface, and slight overheating results in cracks when a bend test is applied. Dark spots may be visible in the fracture of a test-piece or on the surface. It is sometimes stated that overheating may be corrected by long re-heating at the correct temperature, but, with rare exceptions, over-heated metal is fit only for scrap.

The correct solution-treatment temperatures covering a range of standard alloys are reproduced in Table I. It should be noted that material supplied to B.S.S. 2L40, L45, and L47 may include alloys of different chemical composition, necessitating different solution-treatment temperatures. The two types of alloy to each specification are therefore indicated by the letters A and B placed before the specification number. Type A alloys have a copper content exceeding 3·0%, while the copper in Type B alloys is less than 3%.

The normal method of quenching is to plunge the hot materials into cold water, though, owing to variations in contraction from differences in temperature between different parts of the material during the quenching operation, such rapid cooling causes distortion of the work. This can be reduced by controlling the angle at which the metal enters the quenching bath. Other methods employed to reduce distortion are decreasing the cooling rate either by using hot water or oil as the quenching medium, or, for thin material, using an air blast or water spray. All these latter methods, however, have disadvantages; warping of the work may be less, but certain alloys are left liable to intercrystalline corrosion by sea-water, etc.

TABLE I.
CONDITIONS FOR SOLUTION-TREATMENT OF WROUGHT ALUMINIUM ALLOYS.

Specification Numbers.	Form.	*Temperature Range (°C.).	Quenching Medium.
B.S.S.5L3	Sheet and strip	485—505	Water or oil
B.S.S.395	" "	470—490	"
D.T.D.346	" "	510—520	"
D.T.D.356	" "	500—520	"
B.S.S.2L38	Aluminium-coated sheet and coils.	485—505	"
B.S.S.AL47	" " "	500—515	"
B.S.S.BL47	" " "	525—535	"
D.T.D.390	" " "	480—500	"
B.S.S.396	Tubes	470—490	"
B.S.S.3T4	"	485—505	"
D.T.D.220A	"	525—535	"
D.T.D.273	"	480—500	"
D.T.D.450	"	520—530	"
D.T.D.460	"	520—530	"
D.T.D.464	"	495—520	"
D.T.D.520	"	485—505	"
B.S.S.2L37	Rivets	485—505	"
D.T.D.327	"	480—500	"
B.S.S.6L1, 2L39	†Extrusions and forgings	485—505	"
B.S.S.A2L40, AL45	" " "	500—515	"
B.S.S.B2L40, BL45	" " "	525—535	"
D.T.D.364A	" " "	500—520	"
D.T.D.423A, 443	" " "	520—530	"
D.T.D.430A, 410	" " "	515—535	"
B.S.S.2L42	Forgings	510—535	Water
B.S.S.332	"	480—490	Water or oil
B.S.S.333	"	505—520	"
D.T.D.147	Airscrew forgings	470—490	"
D.T.D.150A	" "	485—505	"
D.T.D.184	" "	525—535	Water
D.T.D.246A	Crankcase forgings	515—525	Cool in air
B.S.S.4L25	Bars for forging and forgings	490—525	‡Water or oil
D.T.D.324	Forgings	520—535	Water or oil
D.T.D.363	†Extrusions	450—470	"
B.S.S.177	Bars	480—490	"
B.S.S.478	"	505—515	‡Water or oil

*These temperatures are averages for alloys covered by the Specification, and in cases of doubt reference must be made to the supplier.
† Extrusions = Bars for machining or forging and extruded sections.
‡ Hot water may be used to reduce distortion.

Precipitation and Ageing Treatments

In all age-hardening alloys the strength begins to increase shortly after quenching, but some alloys harden more rapidly and to a greater extent than others. These are known as "natural ageing" alloys, and their full strength is reached after four or five days at normal room temperatures. Hardening is partially prevented by keeping the quenched material just below normal temperature, hence more than five days may be required in cold weather before the maximum strength is obtained. As age-hardening reduces ductility, any appreciable

cold-working must be performed while the metal is still soft. Working of the natural ageing alloys must be completed within two hours of quenching, or, for intricate pressings, within 30 mins. Refrigerators (usually at a temperature of — 6° to — 10° C.) are used to store rivets, sheet and strip up to four days if necessary, to enable more economical runs of work to be performed.

TABLE II.

CONDITIONS FOR PRECIPITATION TREATMENT.

Specification Numbers.	Temperature (°C.).	Times.
Sheet and Strip:		
D.T.D.346	155—175	
D.T.D.356	155—175	
B.S.S.AL47, BL47	170—185	The time required to produce full ageing in a given alloy may be as long as 20 hours or as short as 2 hours, depending upon the temperature of precipitation, the time being a maximum at the lowest temperatures.
Tubes:		
D.T.D.220A	165—175	
D.T.D.460	165—200	
D.T.D.464	155—200	
Forgings and Sections:		
B.S.S.4L27, 478, 533	160	
B.S.S.A21.40, AL45, B2L40, BL45	170—185	
D.T.D.130A, 410	155—175	
D.T.D.184	165—175	The exact treatment required must be obtained from the release note covering the material or from the supplier.
D.T.D.246A	195—205	
D.T.D.324	130—160*	
D.T.D.363	120—140	
D.T.D.364A	155—185	
D.T.D.423A	165—200	

* It is usual to age forgings to this Specification in two stages, namely, up to 10 hours at the temperature given, followed by rough machining and a further period of at least 2 hours at 200° C. ± 10° C.

Some alloys, after solution treatment, harden slightly at normal temperatures, but maximum mechanical properties, particularly proof stress, are not obtained even after long standing. For these alloys a precipitation treatment or "artificial ageing" is necessary. This involves heating the quenched alloy to a temperature between 130° and 200° C. for a definite time, according to the composition of the alloy and the properties required. These double-treatment alloys are thus able to withstand a certain amount of cold work after quenching and ageing at room temperature, without further solution-treatment.

Precipitation-treatment temperatures are summarised in Table II, but the ageing time required depends on the alloy concerned, the temperature selected, and, to some extent, on the type of furnace used. In general, for a specified degree of hardening, high temperatures require shorter times than low temperatures. The time and temperature, therefore, must be selected to give reasonably short time commensurate with accurate control. Heavy or closely packed loads of work in certain types of furnace require a considerable time before the temperature is uniform throughout the mass; in such cases a long heating period is desirable—hence a comparatively low temperature.

The practical information given in the above notes has been abstracted from Information Bulletin No. 3, published by Wrought Light Alloys Development Association, Union Chambers, 83, Temple Row, Birmingham, 2, which is concerned primarily with heat-treatment practice with wrought aluminium alloys, although a useful explanation of the fundamental principle of the heat-treatment of aluminium-base alloys is given in an appendix; The Bulletin also includes four pages of references, covering much important work on the subject. Copies may be obtained on application to the above address.

Temperature Indicating Crayons

AMONG the methods of assessing the temperature of work during operations, that of using indicating crayons provides a very quick means which also possesses a high degree of accuracy; they make available a convenient form of control at a great number of points with the minimum of equipment. These crayons, known as "Tempilsticks," are heat-sensitive compounds which melt at specific temperatures. In the "black range" of temperatures especially they provide a simple means of observing whether certain temperatures exist or have previously been attained. Two methods are recommended:—

(*a*) To make a clear mark on the surface of the material under test, which will melt and for all practical purposes will disappear when the temperature stated on the crayon is attained.

(*b*) To draw the crayon lightly over the surface until a liquid smear remains.

The first method may be used when the work is indirectly heated, whilst the second is applicable to temperature determination where a naked flame is being applied directly to the surface under test.

The crayons are available to indicate the following Fahrenheit temperatures:—125, 150, 175, 200, 225, 250, 275, 300, 350, 400, 450, 500, 550, 600, 650, 700, and they are claimed to have a mean accuracy within 1% of the stated melting-point.

They have many applications in heat-treatment operations, and are especially useful to indicate specific preheating temperatures in welding cast iron, steel, and non-ferrous metals and alloys. In the heat-treatment of heavy and costly castings they can be profitably used to indicate the approach of safe temperature limits, and also assist in avoiding distortion by enabling a check to be kept on the rate of cooling.

Surface temperatures of machinery and motors are readily measured by this means; bearings and insulations can be protected by marking the surface with a crayon and noting when or if a dangerously high temperature is reached. Where the temperature of experiments or processes to be controlled is the same from day to day, a crayon of the desired melting-point will release a more valuable instrument for use elsewhere. These crayons are available from J. M. Steel and Co., Ltd., 36–38, Kingsway, London, W.C. 2.

Zinc Oxide—Maximum Prices Increase

THE Minister of Supply has made the Control of Non-Ferrous Metals (No. 12) (Copper, Lead and Zinc) Order, 1943, which came into force on May 6. This Order increases the maximum prices of red, green and white seal brands of zinc oxide by £3 2s. 6d. per ton in each case, bringing them to the following:—Red Seal, £31 10s. per ton; Green Seal, £33 per ton; White Seal, £34 per ton, for delivery in lots of not less than 2 tons. This high cost has become necessary to cover the higher costs of production and delivery. Enquiries for zinc oxide should be addressed to the Non-Ferrous Metals Control, Grand Hotel, Rugby.

Copies of the Order may be obtained from H.M. Stationery Office, York House, Kingsway, W.C. 2, or through any bookseller, price 1d.

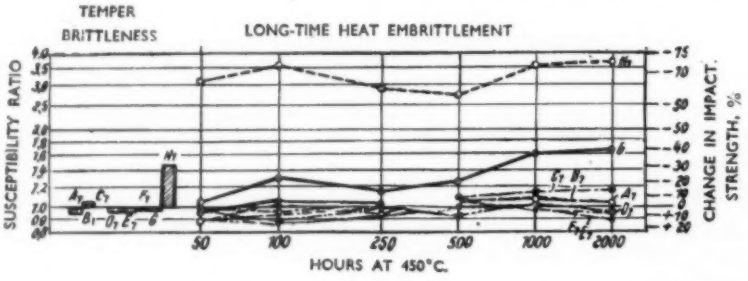
Temper Brittleness and Long-Time Heat Embrittlement

By D. W. Rudorff, M.Inst.F.

A steel is said to be temper brittle if there is a difference between the impact value obtained, respectively, when the steel is quenched in a liquid medium from the tempering temperature, and when it is cooled slowly from this temperature. The susceptibility to temper brittleness is expressed as the ratio between the two values. It is known that certain elements in the composition of a steel contribute to the development of temper brittleness, and efforts have been made to eliminate it by imposing definite limits on the analyses of steels. Although much research work has been carried out on this subject, a completely satisfactory explanation has not yet been found, but a recent investigation, reviewed in this article, is a noteworthy attempt at further clarification of problems of temper brittleness.

CERTAIN low-alloyed steels when slowly cooled from drawing temperature are known to exhibit smaller impact strength than that registered with rapid cooling. A similar tendency towards embrittlement is observed when steels which are sensitive to temper brittleness, are subjected to long-time exposure to temperatures of 400°–550° C. Notwithstanding the fact that a considerable amount of research work was carried out during the last twenty years, a completely satisfactory explanation of these phenomena is still wanting. A noteworthy attempt at a further clarification of these problems is represented by the exhaustive investigations recently made by E. Maurer, O. H. Wilms, and H. Kiessler,[1] who methodically explored the embrittlement characteristics of no fewer than 56 varieties of steels.

These steels were prepared in an acid high-frequency furnace of 50 kilog. capacity. The actual tests for temper brittleness were carried out in the usual manner by establishing the impact strength values after oil quench as well as after slow cooling in infusorial earth at the rate of 2° C. per min. In order to arrive at a convenient standard of comparison, the deterioration in impact strength of the various materials was expressed by the susceptibility ratio. Furthermore, any existing tendency of the steels towards long-time heat embrittlement was ascertained by holding the specimens at 450° C. for periods ranging from 50 to 2,000 hours. By noting initial and final impact strength values, the tendency toward long-time heat embrittlement could be established by expressing the final impact strength in terms of percentage of initial strength.



Steel.	A_1	B_1	C_1	D_1	E_1	F_1	G_1	H_1
C, %	0·13	0·15	0·45	0·38	0·4	0·38	0·42	0·42
Mn, %	0·17	1·12	0·28	0·96	0·18	0·50	0·31	1·08
Ni, %	—	—	—	—	1·7	1·7	5·2	5·1

Fig. 1.—Influence of manganese content upon temper brittleness and long-time heat embrittlement of heat-treated steels.

Influence of the Steel Composition upon Embrittlement Characteristics

It is known that the phosphorus, manganese, chromium and nickel constituents of alloyed steels favour the occurrence of embrittlement. Non-alloyed steels or materials containing no more than mere traces of these elements are practically free of embrittlement. Additions of tungsten and, above all, of molybdenum to steels liable to embrittlement, effect diminution or in some cases even elimination of temper brittleness. The influence of manganese upon the temper brittleness and the long-time heat embrittlement is illustrated by the graph reproduced in Fig. 1, which is drawn on the double-logarithmic scale. Here it is seen that the low carbon steels A_1 and B_1, as well as the medium-carbon steels C_1 and D_1 show no influence of the manganese content upon the impact strength after oil quenching from 850° C. and subsequent tempering. Temper brittleness is exhibited by steel H_1, which contains 1·1% Mn, but not by the similar steel G, in which the manganese content is reduced to 0·31%. However, long-time heat embrittlement tests must be considered as far more valuable and instructive than temper-brittleness investigations, since they serve as a more sensitive and accurate indicator of the tendency of a steel to embrittlement. Taking the case of steel G, exposure for 2,000 hours to 450° C. is seen to result in a decrease in impact strength by approximately 40%; and considering steel H_1, a 30% decrease in impact strength as the result of temper brittleness is contrasted by a 73% decrease in case of the long-time test.

Additions of manganese not exceeding 1% exert their deleterious influence upon impact strength only in steels possessing a tendency towards embrittlement for some

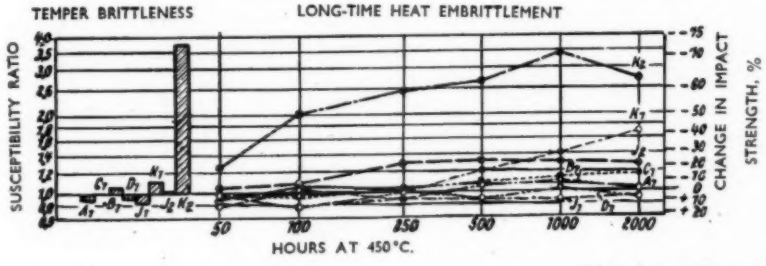
[1] E. Maurer, O. Wilms, and H. Kiessler, "Influence of Phosphorus and of Various Alloying Elements upon Temper Brittleness and Long-time Heat Embrittlement," *Stahl und Eisen*, vol. 62, 1942, p. 81, *et seq.*

other reason—as, for instance, the presence of 5% nickel. Other investigators have pointed out that manganese alone creates temper brittleness only when it is present in amounts exceeding 1·5%. But the exact percentage will of course depend upon the composition of the steel as well as upon the conditions under which the material is examined for temper brittleness. Referring to the behaviour of steel E_1 as compared to that of steel G Fig. 1, shows that nickel contents of 1·5% do not impart any tendency to embrittlement, while with 5% nickel content, long-time heating at critical temperature leads to embrittlement. (It is, of course, well known that pure nickel steels show only little tendency to embrittlement.) To judge from the experiences of other investigators, it would appear that long-time heat embrittlement only occurs with the nickel content exceeding 3%.

Carbon and Chromium Content

Comparatively little attention has been paid in the past to the effect of carbon upon temper brittleness. But it has been noted by Maurer and Hohage[2] that in Cr steels, temper brittleness is governed by the Cr content if the C content remains unchanged, and that vice versa the carbon content becomes the decisive factor if the Cr content is left unchanged. The findings reported in the investigation under review are charted in the graph reproduced in Fig. 2. Here it is seen that the carbon steels A_1 and C_1, with 0·13 and 0·45% C respectively, as well as the 1·0% Mn steels B_1 and D_1 with different carbon contents, are equally free from temper brittleness and long-time heat embrittlement. Steel K_1, containing 0·38% C and 1·0% Cr exhibits a small amount of temper brittleness and a considerable degree of long-time heat embrittlement, but this disappears entirely if the carbon content is lowered to 0·03% (Steel J_1).

Similarly, the increased tendency towards temper brittleness and long-time heat embrittlement of steel K_2, which is characterised by a higher phosphorus percentage, is effectively counteracted by a decrease in carbon content from 0·39% to 0·06% (Steel J_2). This seems to indicate that both temper brittleness and long-time heat embrittlement are somehow directly influenced by the carbon percentage present. Comparing the characteristics of steels D_1 and K_2, it is remarkable that a chromium content of approximately 1·2% proves conducive to embrittlement, while a manganese content of about 1·0% with the carbon content remaining unchanged, does not have any such effect.

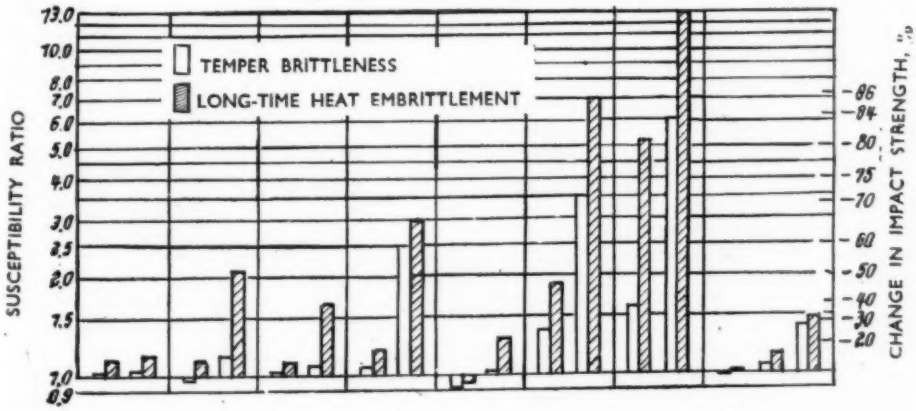


Steel.	A_1	C_1	B_1	D_1	J_1	K_1	J_2	K_2
C, %	0·13	0·45	0·15	0·38	0·03	0·38	0·06	0·39
Mn, %	0·17	0·28	1·12	0·96	0·26	0·49	0·34	0·40
P, %	0·022	0·023	0·025	0·015	0·015	0·012	0·12	0·135
Cr, %	0·1	0·1	0·1	0·1	1·1	1·2	1·1	1·2

Fig. 2.—Influence of carbon content on temper brittleness and long-time embrittlement of heat-treated steels.

Influence of the Phosphorus Content

The effects of increased phosphorus percentages up to 0·129% upon temper brittleness and long-time heat embrittlement of various plain carbon steels and low-alloyed steels are illustrated in Fig. 3. The tempering of these steels was preceded by an oil quench from 950° C. Referring to Fig. 3, it will be seen that the increase in phosphorus content to 0·12% causes a very considerable increase in both kinds of embrittlement in the case of the plain carbon steel C_2 and the manganese steel D_2. The low carbon steel B_2 is also considerably affected, while the plain carbon steel A_2 is hardly affected at all. In the case of the alloyed steels K_1 and L_1, which possess an inherent tendency towards embrittlement, the increased phosphorus percentage

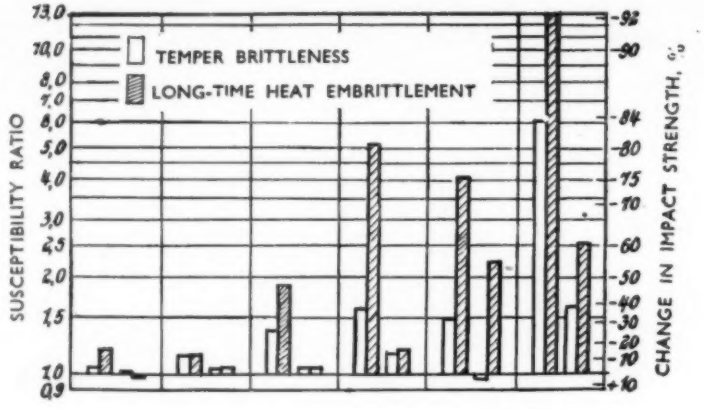


Steel	A_1	A_2	C_1	C_2	B_1	B_2	D_1	D_2	J_1	J_2	K_1	K_2	L_1	L_2	E	E_1	E_2
C, %	0·13	0·14	0·45	0·41	0·15	0·14	0·38	0·36	0·03	0·06	0·38	0·39	0·1	0·37	0·04	0·4	0·42
Mn, %	0·17	0·28	0·28	0·26	1·12	1·04	0·96	0·91	0·26	0·3	0·49	0·40	0·45	0·28	0·06	0·18	0·15
P, %	0·022	0·117	0·023	0·14	0·025	0·121	0·015	0·126	0·015	0·14	0·012	0·14	0·01	0·132	0·155	0·013	0·13
Cr, %	0·1	0·1	0·1	0·1	0·1	0·1	0·1	0·1	1·1	1·18	1·2	1·2	1·3	1·3	0·1	0·1	0·1
Ni, %	—	—	—	—	—	—	—	—	—	—	—	—	1·6	1·8	1·6	1·7	1·7

Fig. 3.—Influence of phosphorus content upon temper brittleness and long-time embrittlement of heat-treated steels.

[2] Mitt. K. Wilh. Inst. Eisenf., vol 2, 1921, p. 91.

leads to an extraordinarily large increase in brittleness (steels K_2 and L_2). As a matter of fact, the Cr-Ni-P steel L_2 shows the greatest embrittlement values of all steels investigated. These results evidence the fact that the embrittling action of phosphorus is fully recognisable only in the case of steels with higher carbon content. From this it must be concluded that only an indirect relationship between phosphorus content and embrittlement exists, which is similar to that experienced with Mn, Cr, and Ni. Moreover, this embrittling action only obtains if a certain phosphorus percentage is exceeded. This conclusion is further affirmed by the behaviour of the steels E, E_1, and E_2 with 1·5% Ni (Fig. 3). Of these three steels, the low-carbon steel E is entirely unaffected by a large phosphorus percentage, while of the two 0.42% carbon steels E_1 and E_2 with 0·013 and 0·126% phosphorus respectively, the latter material shows decidedly inferior impact values. These observations cannot therefore be said to verify Benneck's theory,[3] which postulates that a direct embrittling action of the phosphorus is caused by phosphide precipitation.



Steel.	D_1	D_7	E_1	E_2	K_1	K_2	L_1	L_3	H_1	H_2	L_2	L_{11}
C, % ...	0·38	0·35	0·4	0·38	0·38	0·38	0·4	0·39	0·42	0·49	0·37	0·37
Mn, % ..	0·96	1·0	0·18	0·16	0·49	0·53	0·45	0·67	1·08	1·05	0·8	0·28
P, % ...	0·015	0·018	0·013	0·015	0·012	0·01	0·01	0·022	0·027	0·027	0·13	0·135
Cr, % ..	—	—	—	—	1·2	1·2	1·3	1·4	0·1	0·1	1·2	1·3
Mo, % ..	—	0·33	—	0·33	—	0·39	—	0·37	—	0·46	—	0·33
Ni, % ..	—	—	1·7	1·7	—	—	1·6	1·8	5·1	5·2	1·8	1·7

Fig. 5.—Influence of molybdenum content upon temper brittleness and long-time embrittlement of heat-treated steels.

The inter-relationship between phosphorus percentage and both temper brittleness and long-time heat embrittlement of a 1·0% Mn steel is shown in the graph reproduced in Fig. 4. Referring to this graph, it will be seen that a high degree of embrittlement obtains with a phosphorus content of 0·09%. Beyond this percentage, a much slower rise in brittleness with the phosphorus content takes place. It is also interesting to note that above the aforementioned percentage, the temper brittleness and long-time heat embrittlement curves run practically in parallel.

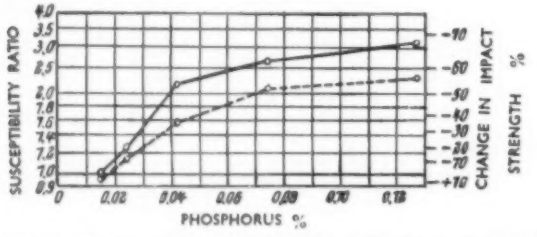


Fig. 4.—Influence of increased phosphorus percentage upon temper brittleness and long-time embrittlement of heat-treated steel containing 0.4%C, 0.2%Si, and 1.0%Mn.

The Molybdenum Content

The beneficial effect exerted by additions of molybdenum upon the brittleness of various types of steel is particularly well exemplified by the comparative data presented in Fig. 5. This graph shows that the addition of molybdenum to steels containing 1·0% Mn or 1·7% Ni almost completely suppresses the small tendency toward embrittlement inherent in them. This also applies to the 1·2% Cr steels K_1 and K_2. However, the beneficial effect of molybdenum additions is much smaller in the case of the Cr-Ni steels L_1 and L_3, and L_2 and L_{11}, and the improvement obtained is likewise very small in the case of the 5% Ni steels H_1 and H_2. The investigators therefore decided to ascertain whether a more radical

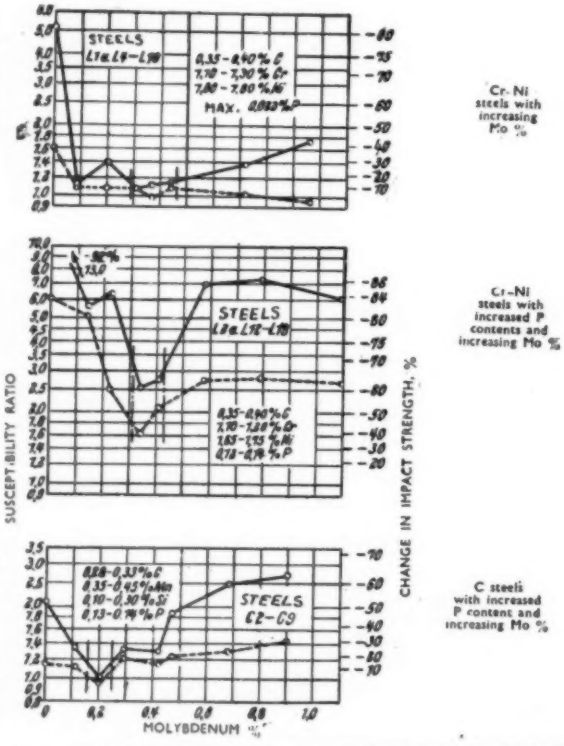
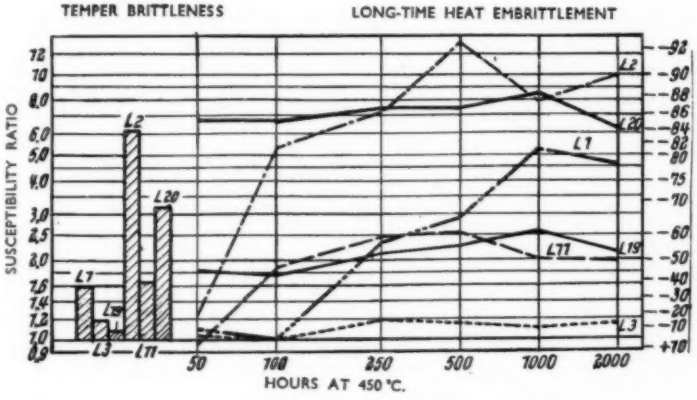

STEELS
Cr-Ni steels with increasing Mo %
Cr-Ni steels with increased P contents and increasing Mo %
C steels with increased P content and increasing Mo %
SUSCEPTIBILITY RATIO
CHANGE IN IMPACT STRENGTH, %
MOLYBDENUM %
TEMPER BRITTLENESS
LONG-TIME HEAT EMBRITTLEMENT

Fig. 6.—Influence of increasing molybdenum content upon temper and long-time heat embrittlement of heat-treated steels with and without increased phosphorus content.

[3] *Arch. Eisenh.*, Wesen, vol. 9, 1935-36, p. 147.

improvement can be obtained with larger molybdenum additions. For this purpose a series of tests was run in which molybdenum additions varying from 0·1–1·0% were made to (1) the class of steels C_2-C_9, containing an increased amount of phosphorus (see Fig. 6); (2) the class of Cr-Ni steels L_2 and L_{12}-L_{18} with increased phosphorus percentage (Fig. 6); and (3) the class of Cr-Ni steels L_1 and L_4-L_{10} with usual phosphorus contents (Fig. 6). The test results incorporated in the aforementioned graph show that in all steels investigated the inherent tendency towards embrittlement is progressively reduced by the addition of varying molybdenum percentages until a certain molybdenum percentage is reached, above which the tendency towards embrittlement again increases.



Steel.	L_1	L_3	L_{19}	L_2	L_{11}	L_{20}
P, %	0·01	0·022	0·018	0·132	0·135	0·032
Mo, %	—	0·37	—	—	0·33	—
Nb, %	—	—	0·25	—	—	0·2

Fig. 7.—Comparison of the influence of Mo, Nb, and P contents upon temper brittleness and long-time heat-embrittlement of heat-treated steels with approximately 0.4%C, 13%Cr, and 1.7%Ni.

This phenomenon is especially pronounced with the steels containing increased phosphorus percentage. This fact leads to the important conclusion that there exists for each steel a rather low optimum molybdenum percentage which must not be exceeded if highest impact strength characteristics of the material are aimed at.

The Influence of Niobium

A few investigations were also made regarding the influence of niobium upon the embrittlement characteristics of a 0·4% carbon, 1·3% chromium, and 1·7% nickel steel with varying phosphorus content. As Fig. 7 shows, an addition of 0·25% Nb (steel L_{19}) proves slightly more effective in reducing temper brittleness than the addition of 0·37% Mo (steel L_3). However, the reverse holds true in the long-time heat embrittlement test. In the case of the high-phosphorus steels L_2, L_{11}, and L_{20} the result is still more disappionting insofar as the Nb-alloyed steel L_{20} shows practically the same long-time heat embrittlement found with the high-phosphorus steel L_2. It must therefore be concluded that the embrittlement-impeding action of niobium is in no way comparable to that of molybdenum.

Influence of Heat-treatment upon Embrittlement Tendency

A special series of tests was made to ascertain the influence of quenching temperature and speed, and of heat-treatment below the A_1 point. For this purpose two specimens each of a number of steels with different tendency towards embrittlement were quenched in water from 850° and 950° C. respectively. All specimens were then tempered at 650° C. followed by cooling in oil or infusorial earth. All steels quenched from the higher temperature exhibited a definite increase in susceptibility as well as in long-time heat embrittlement. A particularly good insight into the relationship between quenching temperature and temper brittleness is afforded by the graph reproduced in Fig. 8, which refers to a

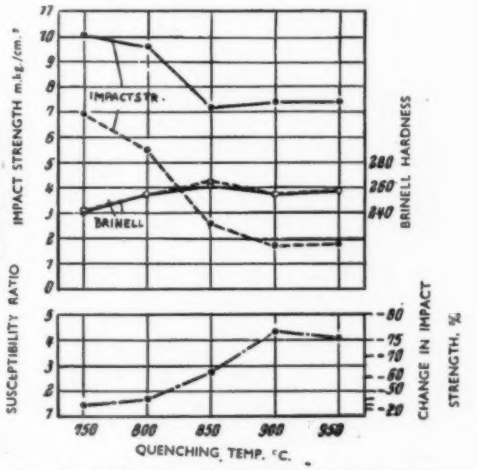


Fig. 8.—Influence of quenching temperatures upon temper brittleness of heat-treated steel with approximately 0.36%C, 0.91%Mn, and 0.126%P.

Mn-P steel. The lowest quenching temperature employed was 750° C. This lies above the Ac_1 point of 737° C., but it is still below the Ac_3 point of 797°C. The graph shows that temper brittleness already prevails in the range from 750°–800° C., but it progressively increases when raising the quenching temperature from 800°–900° C.

That the quenching speed is likewise of influence upon the brittleness of the tempered material was evidenced by the observation that the aforementioned Mn steel D_1, the high-phosphorus steel A_2, and the Ni steel E_1 exhibited a greater susceptibility with air quenching than with oil quenching.

Heat-treatment Below the Ac_1 Point

It was already pointed out that the long-time heat embrittlement is a better measure of the embrittlement tendency of a steel than the temper brittleness. It was, however, observed that the embrittlement effected by heating at 450° C. reached the corresponding temper brittleness value only after 50–100 hours, or after even longer periods. From this it was deduced that in the cases concerned, the temperature of 450° C. did not constitute the temperature of maximum embrittlement within the critical temperature range. It was therefore decided to ascertain the long-time heat embrittlement of the 0·36% C, 0·91% Mn, and 0 126% P steel D_2 in

heat-treated state over the temperature range of 300°–650° C. Subsequent to each test run, one specimen was cooled in water and another specimen was slowly cooled by embedding it in infusorial earth. This measure was resorted to in order to ascertain both the temperature of greatest long-time heat embrittlement and the influence of the cooling speed.

Referring to Fig. 9, the left-hand side of the diagram shows the impact strength both in tough annealed condition and in temper brittle condition. The juxtaposed curve of impact strength indicates that the critical temperature range extends from 400°–600° C., a minimum of impact strength occurring with 500° C. temperature. Up to this temperature, the cooling speed is without influence upon the impact strength, but this changes in the recovery range extending from 500°–650° C. After holding the steel at 650° C. for 100 hours, the temper brittleness was found to amount to only one-third of that observed immediately after tempering. This seems to confirm earlier observations made by Houdremont and Schrader[4] as well as by Reichel,[5] to the effect that prolonged holding at tempering temperature diminishes temper brittleness. The graph also shows that in the range from 400°–550° C., the process of embrittlement is in no way connected with the hardness, since the latter is seen to remain constant over this temperature range The rapid fall in Brinell hardness at temperatures exceeding 600° C. of course is due to the tempering effect. The beneficial influence of long-time tempering upon susceptibility and long-time heat embrittlement is well exemplified by the data found with the steel K_2 containing 0·39% C, 0·4% Mn, 0·135% P, and 1·2% Cr. These findings are charted in Fig. 10.

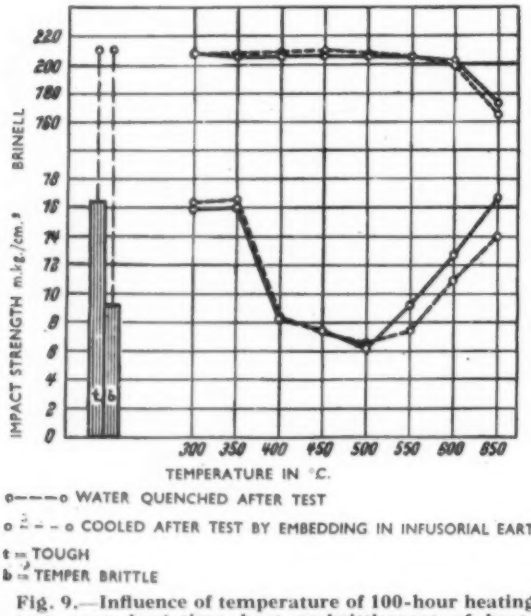


Fig. 9.—Influence of temperature of 100-hour heating test upon long-time heat embrittlement of heat-treated steel with 0.36%C, 0.91%Mn, and 0.126%P.

Summarising their findings, the investigators express the opinion that nickel, manganese, chromium, and phosphorus have an indirect influence upon the susceptibility of the steels investigated, the influence of phosphorus increasing progressively with increasing percentage. Plain carbon steels and steels with low percentages of the aforementioned elements, show no tendency to embrittlement if their carbon percentage is small. A certain minimum carbon percentage therefore is a pre-requisite to embrittlement. Susceptibility to embrittlement is increased by the employment of high quenching temperatures. Concerning the influence of the cooling speed employed in the hardening process, a slow cooling rate will lead to an increase in the susceptibility of steels with little tendency towards embrittlement, and to a decrease in susceptibility of steels normally showing a pronounced tendency towards embrittlement. Long-time heating at temperatures below the A_1 point, but above the critical temperature range of embrittlement, serves to decrease subsequent tendencies to embrittlement. The latter conclusion is quite in agreement with the statements of other investigators.

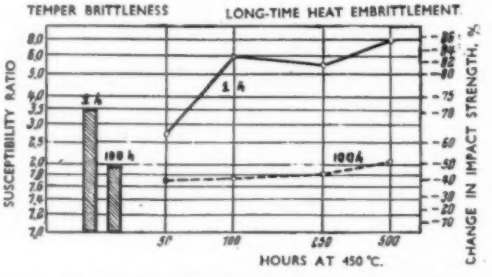


Fig. 10.—Influence of long-time tempering at 650° C. upon temper brittleness and long-time heat embrittlement of heat-treated steel with 0.39%C, 0.13%P, and 12 %Cr.

[4] *Arch. Eisenh.*, Wesen, vol. 7, 1933-34, p. 49.
[5] Doctor Thesis, Freiberg, 1933.

New Swedish Institute for Metallographical Research

As in many other countries, research is the watchword in Sweden to-day. Special attention is devoted to fields in which Swedish products are used as raw materials. A special committee appointed by the Government to investigate the need of new research institutions has recently submitted to the Government a proposal for the establishment of a new institute for iron and steel research, called the Metallographical Institute.

In its proposal, the Committee points out that the Swedish iron and metal industry is a typical "quality" industry with a position to uphold which requires untiring efforts to keep abreast with progress in this field. Sweden's export of iron and steel products is mainly dependent upon the country's ability to produce high-quality, so-called "fine" steel and steel products. To keep these products up to the highest standard and yet to be able to produce them at competitive prices will demand unremitting research.

It is proposed to erect the new institute near the Royal Technical University of Stockholm.

The Structure of Rolled and Annealed Aluminium as Revealed by X-Rays

By E. E. Spillett, B.Sc.

X-ray investigations on the structure of rolled and annealed aluminium sheet and strip were begun in the laboratories of the British Aluminium Co., Ltd., in 1928, and were continued until the outbreak of war. The initial survey of the structural changes associated with cold rolling and annealing was begun by the author's colleague, Mr. H. W. L. Phillips, and the author proceeded with the study of these problems, a comprehensive account of which is given in a paper published in the Journal of the Institute of Metals and is presented here in a substantially abridged form.*

THE process of fabricating aluminium sheet and strip takes place in two main stages. The cast ingot or billet is first hot-rolled to a convenient thickness, and is then cold-rolled, either in sheet mills, with light passes, or in strip mills. In the latter the passes are heavier, and some tension is applied by means of the coiling mechanism. If soft-temper material is required, the product is finally annealed; for intermediate tempers anneals are given in process.

The structure of the metal at the end of the hot-rolling stage is not homogeneous, and usually consists of recrystallised layers with a cold-worked centre core (see Fig. 1). Sometimes a fairly sharp line of demarcation between the two regions can be detected on etched micro-sections, and the extent of recrystallisation depends chiefly upon the thermal conditions and the amount of working during the hot-rolling process. Transmission X-ray diffraction patterns from various layers isolated by suitable etching, reveal that whereas the surface is completely recrystallised with almost random orientation, the centre is cold-worked with pronounced preferred orientation.

During cold-rolling the crystals are distorted and broken into fragments; these fragments tend to align themselves in one of the more important crystallographic directions, but the alignment is never perfect, and appears to be statistical, the histograms being of the single-humped type. If a polycrystalline aggregate, in which the crystal orientation is fully random, is irradiated with monochromatic radiation, the crystals give their diffraction spectra as isolated spots uniformly distributed on the Debye-Scherrer rings; and when the grain size is very fine these spots merge into continuous rings. When recrystallised metal undergoes deformation, the crystals are at first distorted and subsequently fragmented, giving rise to a blurring of the spots on the X-ray diffraction patterns, followed by circumferential extensions of the spots, so that continuous rings are formed. Increased deformation induces a preferential alignment of the crystallites, causing the Debye-Scherrer rings to break up into arcs which contract in length as the orientation approaches perfection.

When cold-rolled aluminium is heated sufficiently, mechanical softening occurs, and at a later stage recrystallisation takes place; the latter process can be followed by X-ray analysis. The new crystals appear as sharp isolated spots on the Debye-Scherrer rings, and the spots increase in number and size as the recrystallisation proceeds to completion; at this stage the original worked structure is replaced entirely.

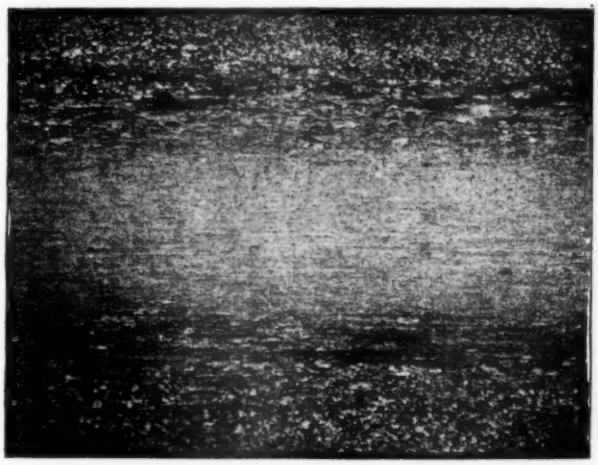


Fig. 1.—Structure of hot-rolled blank. x 12. Etch: NaOH, HF.

Structure of Sheet and Strip Cold-rolled without Intermediate Annealing

Reference has been made to the heterogeneous nature of a hot-rolled blank of commercial purity aluminium at the completion of hot-rolling, and whilst Fig. 1 shows an etched section, Fig. 2 reproduces transmission X-ray diffraction patterns taken at various positions throughout the thickness of the blank. The marked preferred orientation at the centre approaches that most frequently associated with rolled metals of the face-centred cubic type—namely, the crystallites tend to become aligned with a (112) direction and a (110) direction normal to the rolling planes. Fig. 3 shows diagrammatically a crystal oriented in this manner, one set being the mirror image of the other by reflection in a plane of symmetry perpendicular to the plane of rolling and running longitudinally. The alignment is not perfect; if it were, none of the planes registered on the pattern would be suitably oriented to reflect according to the Bragg law, and no pattern would be obtained. It is estimated that

* April, 1943 issue.

in the longitudinal direction the variation of individual fragments from the mean position reaches 10°, whereas in the other directions the variations are greater.

No signs of recrystallisation are evident in the central layers, yet in intermediate layers between surface and centre a partially recrystallised pattern is obtained; the preferred orientation of the unrecrystallised fragments is not the same as that obtaining in the centre, but approximates to a (110) direction in the rolling and transverse directions.

When such material is cold-rolled, the difference

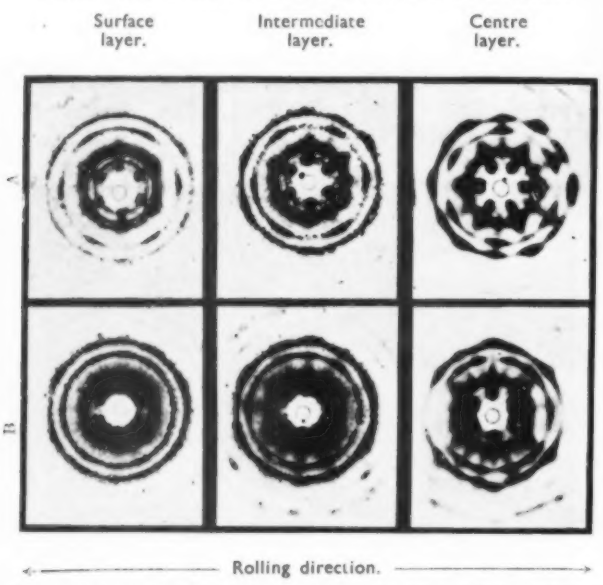


Fig. 2.—X-ray transmission patterns on hot-rolled blank.
A.—X-ray beam in transverse direction.
B.—X-ray beam normal to rolling plane.

between the surface and centre observed on micro-sections gradually disappears, and with reductions in excess of 40% such sections show the typical striated structure associated with cold-rolled metals. There is still, however, a pronounced difference in structure

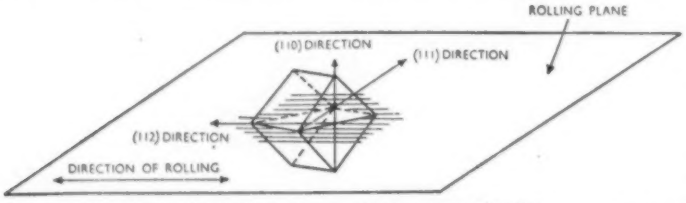


Fig. 3.—Crystal orientation shown diagrammatically.

between surface and centre, as revealed by transmission X-ray examination of isolated layers. At the centre the preferred orientation is unchanged, but in the surface layers the cold-working first causes fragmentation, the crystallites being first oriented at random; and with further working preferred orientation develops. This orientation, however, never becomes so preferred as that at the centre, and even after heavy reductions (90% and over) from the hot-rolled ingot; there is still a marked difference between surface and centre. Some X-ray patterns which are reproduced in Fig. 4 show the structure at the centre and surface of strip of 99·6% purity aluminium rolled from hot-rolled blank with reductions of 40 and 90%.

The orientation as estimated from transmission patterns of isolated layers gradually becomes more preferred from the surface to the centre of the rolled sample: there is no sharp dividing line.

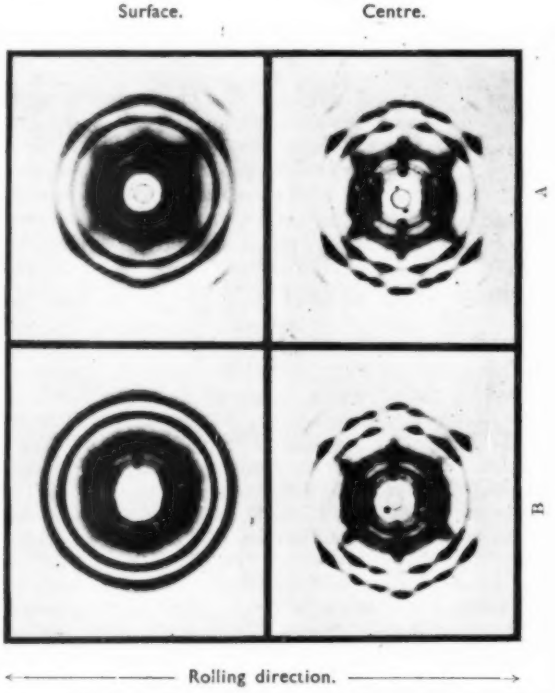


Fig. 4.—Difference in structure between surface and centre of hard-rolled strip.
A.—90% reduction.
B.—40% reduction.
X-ray beam normal to rolling plane.

With increased working the material becomes harder and directional differences in the mechanical properties are developed. In one experiment it was found that after light reductions from hot-rolled blank there was no appreciable difference in tensile strength in various directions, yet with increased reductions, as the orientation became more preferred, the tensile strength increased and maximum values were obtained in the longitudinal and transverse directions, and minimum values were obtained at 45° to these directions.

There is strong evidence to suggest that two different types of longitudinal fibring develop in the surface layers of rolled aluminium, one similar to that obtained at the centre, whereas the other has a (110) direction to the rolling direction. The extent of the development of the different types apparently depends upon the method of rolling. In strip rolling, when the metal is under tension, owing to the action of the rolling mechanism, and comparatively large reductions per pass are given, there is a greater tendency for the (112) type to develop; whereas with sheet rolling, where relatively light passes are employed, there is a preferential development of the (110) type. The development of (110) fibring is in agreement

with the usually accepted compression structure, suggesting that in sheet rolling the deforming forces are probably more of a compressive nature. This difference between sheet and strip-rolled aluminium can sometimes be detected on transmission X-ray diffraction patterns of the complete thickness, and typical cases are given in Fig. 5. The pronounced maxima at 0° and 180° on the (220) Debye-Scherrer ring are due to the (110) fibring.

The difference can be clearly seen on X-ray diffraction patterns obtained by glancing incidence with the X-ray beam in the transverse direction; the orientation changes from (110) fibring at the surface to the (112) fibring at the centre in the case of sheet-rolled aluminium. With strip it appears that the (112) type only is developed.

Abnormalities are sometimes experienced, in particular in the surface layers, where tilted fibring may occur: this is indicated by an asymmetrical transmission pattern, the maxima approaching one another in one half of the pattern and separating in the other. Typical examples are given in Fig. 6, which are from the surface layers of a hard-rolled strip sample. The greater tendency for tilted fibring to occur in the surface layers may be visualised by considering the passage of a sheet through the rolls. In the centre the path of the crystals remains horizontal and they tend to align themselves with the fibre axis horizontal; towards the surfaces, however, the crystals follow a curved path, and the tendency will be for the fibring to take place parallel to the direction of flow at any particular moment. In the surface layers flow only becomes horizontal at the moment of exit, and thus there is less opportunity for horizontal realignment.

Structure of Sheet and Strip Cold Rolled after Intermediate Annealing

When annealed aluminium is cold-rolled, the crystals become distorted and elongated as the reduction is increased, although even after heavy reductions it is still possible, on etched micro-sections, to identify the original crystals. The first effect observed on the X-ray transmission patterns is a blurring of the spots on the Debye-Scherrer rings, and in a sheet 0·032 in. thick the effect first appears with reductions in the region of 1–2%. This limit is only approximate, as it is difficult to obtain accurately by rolling such small reductions in thickness.

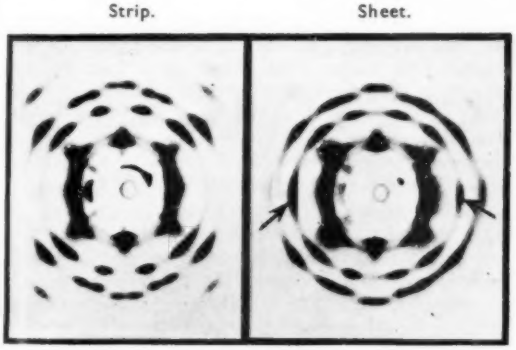


Fig. 5.—Difference between sheet and strip rolling as shown by X-ray transmission patterns. X-ray beam normal to rolling plane.

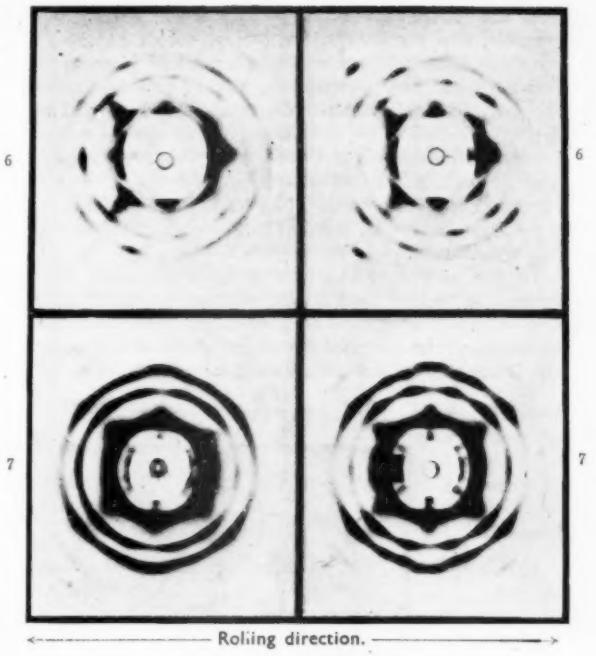


Fig. 6.—Tilted fibring.

Fig. 7.—Strip rolled with 90% reduction from as-rolled (right) and annealed (left) hot-rolled blank. X-ray beam normal to plane of rolling.

A similar effect is obtained if the working is applied by stretching, when more accurate control of the deformation can be obtained. In this case the first effect on the X-ray pattern appears when the sample is extended permanently between 1 and 2%. The interesting feature is that the first signs of deformation on such patterns are general, and not localised as would be expected on the grounds that certain crystals—those oriented in the directions most favourable to slip—are likely to be deformed before, and to a greater extent than others less favourably situated.

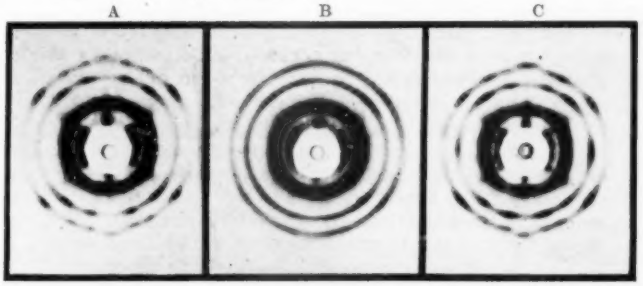


Fig. 8.—X-ray transmission patterns showing effect of incomplete intermediate anneal. X-ray beam normal to plane of rolling.

In an endeavour to follow the initial process of deformation more closely, the author's colleague, Mr. H. W. L. Phillips, prepared some large grains of the order of 3 mm. in diameter in a strip of 99·9% purity aluminium by the standard method of applying a small deformation and subsequently annealing at a high temperature. Transmission X-ray patterns were obtained

from the centre and boundary of two selected individual crystals, and similar patterns were taken at the same positions after various rolling passes had been applied. It was found that despite the use of filtered radiation the initial patterns had to be treated as Laue patterns, since unfiltered white radiation gave images of sufficient intensity to be recorded even from planes of relatively high indices. After further working, however, the monochromatic pattern was developed as in the case of polycrystalline material, and the white radiation played no important role.

In the centre of the first crystal examined, it was surprising to find that initial rotation occurred without appreciable " warping " or buckling of the crystal planes, although at the boundary rotation and warping occurred simultaneously, and at an earlier stage in the rolling than at the centre of the crystal. It appeared that the neighbouring crystal appreciably affected the behaviour at the boundary of the crystal under examination. In the second crystal, selected from a different area of the strip, rotation and warping occurred simultaneously at the centre, whereas at the boundary the deformation lagged behind that at the centre, and it appeared that the neighbouring crystal was deforming in such a manner as to restrict the deformation of the boundary of the crystal under examination. In both crystals this type of deformation appeared to continue until a certain amount of fragmentation (as shown by a circumferential spreading of the spots and decrease in radial streakiness) had set in. It was not possible to correlate the start of fragmentation or the cessation of rotation with any particular reduction, as it varied considerably in the different areas examined. These experiments thus revealed several interesting features: in particular, it may be possible for rotation to occur without appreciable buckling, and strain gradients may be developed across individual crystals. It seems that individual crystals may deform differently, and that deformation can be greater at the boundary than in the body of the crystal or vice versa.

When annealed polycrystalline material is cold-rolled, the crystals become fragmented as the rolling proceeded, and the resulting crystallites increasingly tend to be aligned in a manner similar to those in sheet and strip rolled without intermediate annealing. Fig. 9 reproduces patterns of two series of strip-rolled aluminium of two different purities (0·18% silicon, 0·61% iron, and 0·15% silicon, 0·33% iron). The patterns are from samples all 0·012 in. thick, rolled with various reductions (33 to 92%) from annealed aluminium, and also from samples rolled without an intermediate anneal (94% reduction from hot-rolled blank). The development of preferred orientation with increased reduction is clearly shown, and the patterns also illustrate the effect of purity. For any given reduction, the orientation developed is not so preferred with the higher iron content, and the difference is more striking on the original patterns. Thus, it is not possible to estimate from an X-ray pattern, with any degree of certainty, the post-anneal reduction, unless the purity is known and appropriate standards are available for comparison. Fig. 9 also shows that even with heavy reductions from the intermediate anneal the preferred orientation never becomes so preferred as in material rolled direct from hot-rolled blank.

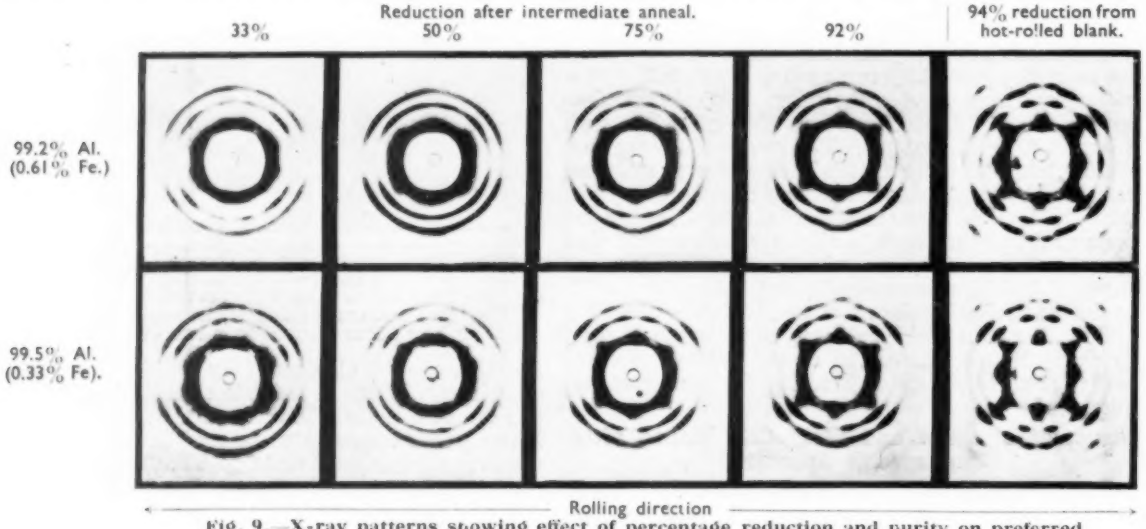


Fig. 9.—X-ray patterns showing effect of percentage reduction and purity on preferred orientation. X-ray beam normal to rolling plane.

It is generally found that if the intermediate anneal is imposed sufficiently remote from the finishing gauge, there is no appreciable difference in tensile strength or hardness between hard-rolled and intermediately annealed material, although there is a marked difference in X-ray pattern. To illustrate this, there are reproduced in Fig. 7 patterns from 99·3% aluminium strip rolled from annealed and unannealed hot-rolled blank, the tensile strengths being 11·6 and 11·5 tons/sq. in. respectively.

X-ray examination may therefore be used to detect whether an intermediate anneal has been applied, in cases where it is impossible to determine this from mechanical tests. It is not infallible, however, since if the anneal has not been sufficient to bring about recrystallisation and to remove the original orientation, the X-ray pattern of the final sample will still show strongly preferred orientation, and it will be impossible to say whether an intermediate anneal has been given or not. This is illustrated in Fig. 8, which reproduces

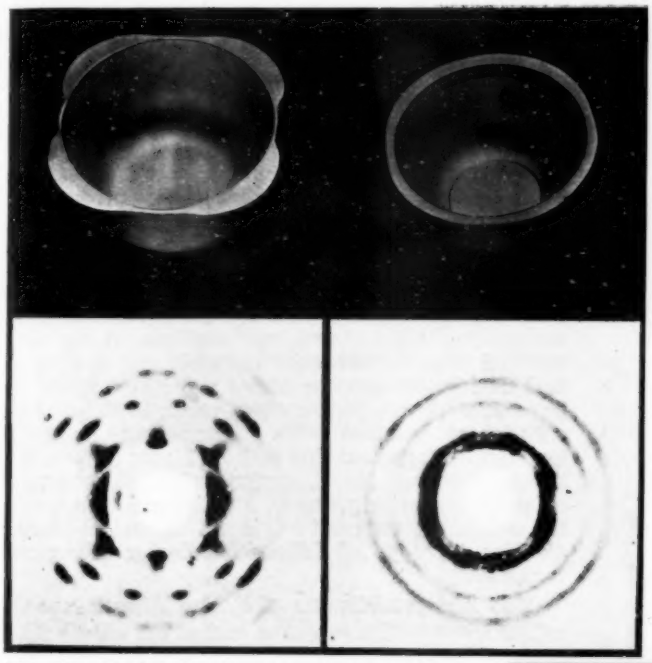


Fig. 10.—Earing developed during pressing, owing to incomplete recrystallisation at intermediate anneal.

X-ray transmission patterns from three samples of rolled aluminium of the same thickness and purity (99·3%) prepared as follows :—

(A) Rolled direct from hot-rolled blank. Sclerescope hardness 18.

(B) Completely recrystallised at an intermediate anneal, followed by 30% reduction. Scleroscope hardness 12.

(C) Partially recrystallised at an intermediate anneal, followed by 30% reduction. Scleroscope hardness 12.

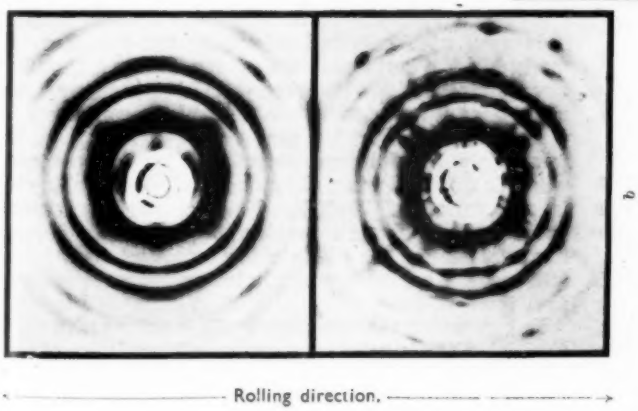


Fig. 11.—Effect of grain size at intermediate anneal on final structure of 99.9% aluminium strip. Specimens intermediately annealed at (a) 360° C. and (b) 500° C., followed by 75% reduction.
X-ray beam normal to the plane of rolling.

It will be seen that, as between A and C, the X-ray examination is insufficient to show whether the material has been intermediately annealed or not, but the required information is furnished by the hardness figures. As between B and C, however, the X-ray examination shows that the former material has been completely recrystallised at the intermediate anneal, whereas the hardness test does not. Thus, both tests must be used in conjunction.

As preferred orientation is an indication of directional properties, the X-ray method is invaluable in connection with the examination of material employed for deep drawing or other forming operations. Fig. 10 illustrates an interesting example of this ; one pressing shows marked ears, whilst the other exhibits no appreciable directionality. X-ray patterns taken from the undeformed metal at the base reveal a marked difference in structure, although hardness tests showed no difference. The information thus suggests that the material from which the defective pressing

Fig. 12.—99.4% aluminium strip intermediately annealed at (a) 500° C. and (b) 360° C., followed by 75° reduction.
X-ray beam normal to plane of rolling.

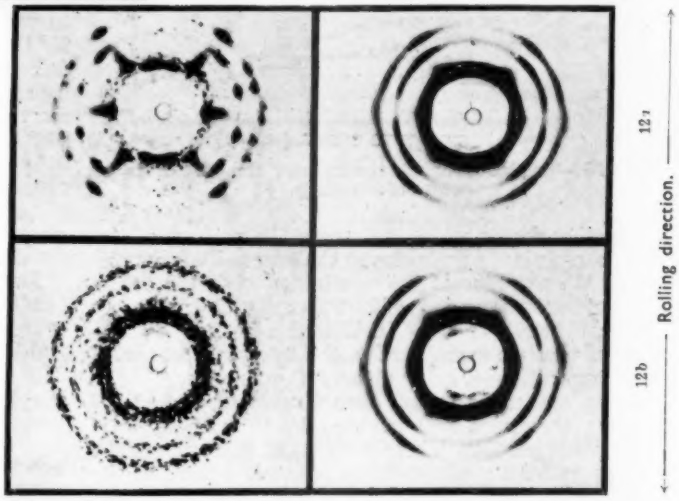


Fig. 14.—Difference between centre (a) and surface (b) of hard-rolled strip after heating to 300° C.
X-ray beam normal to plane of rolling.

was prepared was not completely recrystallised at the intermediate anneal, although a similar effect might also be obtained by partially annealing hard-rolled material.

The above illustrations serve to show that whereas valuable information may be derived from X-ray examination which cannot be obtained by other means, it is not always possible to derive the full history from X-ray examination alone. The results of such examination should be used in conjunction with other data to avoid the possibility of erroneous conclusions. The above examples also show that the extent of preferred

orientation cannot always be correlated with the amount of working as estimated from mechanical tests.

Provided the reduction after the process anneal exceeds 60%, so that the preferred orientation is sufficiently well developed to be identified, it can be seen that a difference exists between sheet and strip as regards the type of preferred orientation. This difference is similar to that noted in hard-rolled material. The difference in the extent of fibring between the surface and centre is not so marked as in hard-rolled material, even after heavy reductions. There is no doubt that the marked preferred orientation existing in the centre of hard-rolled material is a legacy from the hot-rolling treatment, which may be removed by suitable intermediate annealing.

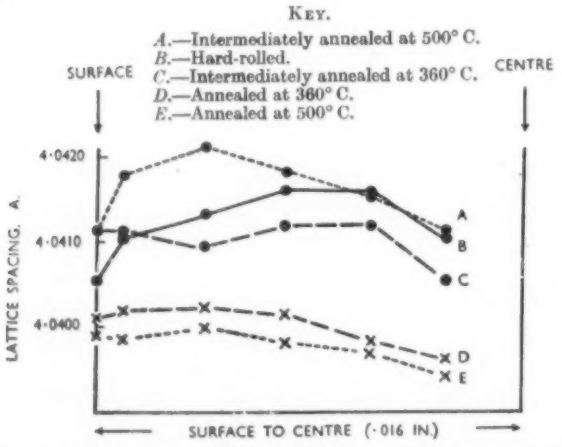


Fig. 13.—Lattice spacing throughout the thickness of aluminium strip.

Another factor which influences the pattern of the final sample is the grain-size at the intermediate anneal. Fig. 11 reproduces the patterns obtained after 75% reduction from strip of 99·9% purity intermediately annealed at 360° and 500° C. With this purity, the grain-size of the two samples after the intermediate anneal was very different, being about 500 grains/mm.² in the former, and probably less than 50 grains/mm.² in the latter; the grains were too irregular to make an exact evaluation possible. There is a marked difference in the X-ray patterns; where the grains were coarser at the intermediate stage, the preferred orientation developed is less regular. With 99·4% purity metal, however, there was little difference in the grain-size after annealing at 360° and at 500° C., and, as will be seen from Fig. 12 (Plate X), there is no difference in the patterns obtained after 75% cold reduction, following the intermediate anneal.

The Lattice Spacing of Cold-Rolled Aluminium

Some of the energy absorbed during cold-rolling is used in the fragmentation and orientation of the crystals and some is stored in the crystals in the form of potential energy which can be estimated from lattice distortion. Lattice-spacing values have been derived from the reflections from the (333) and (511) families of planes, using a precision camera of the self-focusing type. The results show a fair amount of scatter, but if an adequate number of tests be made the results can be analysed statistically to investigate the significance of any variations.

The full line in Fig. 13 illustrates the variation in parameter which is normally experienced throughout the thickness of cold-rolled aluminium strip (rolled without intermediate annealing). The curve represents the mean results obtained from duplicate tests on each of six specimens (99·4% aluminium), and shows that whilst the lattice spacing is relatively low at the surface of the strip, the spacing increases below the surface, rises to a maximum, and then decreases again towards the centre. Reproduced on the same illustration are values obtained in a similar manner for intermediately annealed material of the same purity and thickness (0.032 in.). As expected, the spacing is lower and more uniform in annealed material than in cold-rolled material, and it appears that the high-temperature anneal may give a relatively low value. By considering these values in conjunction with those obtained with intermediately annealed material, it is apparent that with the higher temperature intermediate anneal the distortion during the post-anneal reduction is relatively greater. It follows, therefore, that the annealing temperature is an important factor affecting the type of deformation during subsequent working.

The Recrystallisation of Rolled Aluminium

This investigation has been carried out chiefly by a study of X-ray transmission patterns of samples heated to various temperatures, usually at intervals of 10° or 25° C. Dealing with the recrystallisation of lightly deformed samples of 99·4% purity aluminium, prepared by stretching dead soft material, it is found that the recrystallisation temperature is lowered from 600° to 350° C. as the extension before heating is increased from 0·1 to 20%. A careful examination of the X-ray patterns reveals that no sharpening of the blurred spots occurs either before or during recrystallisation; the new crystals give rise to well-defined isolated spots randomly distributed on the Debye Scherrer-rings.

Two series of strip-rolled samples were prepared; a hard-rolled series rolled direct from hot-rolled blank with reductions varying from 28 to 94% (0·144 in.–0·012 in. thick), and a series intermediately annealed at 360° C. all finished at 0·012 in. thick, with reductions varying from 33 to 92%. Two compositions of commercial purity aluminium were used, differing chiefly in iron content—viz., 99·5% (silicon 0·15, iron 0·33%), and 99·2% (silicon 0·18, iron 0·61%).

With the metal containing the higher percentage of iron, the temperature at which recrystallsation occurs is lowered with increased reduction, and this is most marked in the hard-rolled series. It was found that for any given reduction a lower recrystallisation temperature is obtained if an intermediate anneal has been given.

The series with higher purity showed that there is apparently a minimum recrystallisation temperature which is obtained when the reduction by cold-rolling is in the region of 75–80%. With greater or smaller reductions a higher temperature is required to induce recrystallisation. As with the lower-purity material, an intermediate anneal apparently lowers the recrystallisation temperature.

In comparing the two purities, it is noteworthy that the higher purity material shows greater variation of the recrystallisation temperature with the amount of cold-rolling, and that it tends to recrystallise at a higher

temperature for any given reduction. These particular differences are probably due to the impurity constituents acting as subsidiary nuclei for recrystallisation, since it is known that they may act as nuclei and more are present in the case of the lower-purity metal.

It is often difficult to determine the temperature necessary for the attainment of complete recrystallisation from X-ray analysis, and this difficulty is particularly marked in aluminium with a relatively high iron content, owing to the persistence of the original preferred orientation after recrystallisation if heavy pre-anneal reductions are employed. With relatively low iron contents, complete recrystallisation with random orientation is usually obtained with or without an intermediate anneal, and irrespective of the pre-anneal reduction, but with high iron contents complete recrystallisation with random orientation can only be ensured by giving an intermediate anneal and by limiting the final reduction to about 75%.

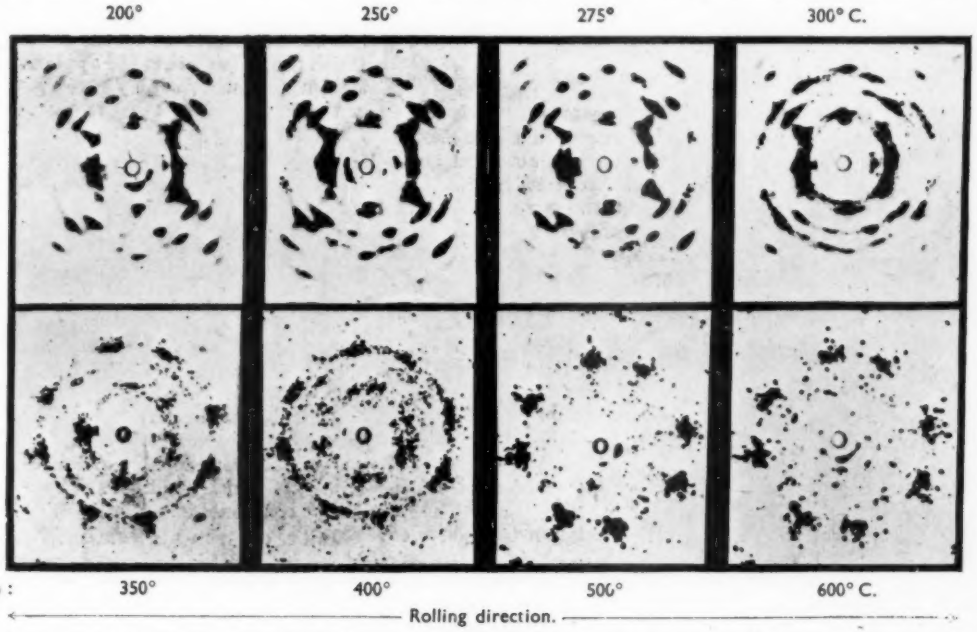


Fig. 15. — X-ray patterns showing development of preferred orientation in super-purity aluminium. X-ray beam normal to plane of rolling.

In dealing with the recrystallisation throughout the thickness of rolled aluminium, it is invariably found that if no intermediate anneal is given recrystallisation first begins in the surface layers, and an increase of the order of 20° C. is required to promote recrystallisation in the centre. Furthermore, the surface layers are completely recrystallised at a temperature about 30° C. lower than that required to bring about complete recrystallisation at the centre. Fig. 14 reproduces X-ray transmission patterns from the surface and centre layers of the strip with 90% reduction illustrated in Fig. 4, after heating to 300° C. ; the patterns illustrate the sluggish nature of the recrystallisation in the central layer. If, however, an intermediate anneal at 360° C. is given at four times the finishing thickness (i.e., 75% reduction), then in subsequent annealing recrystallisation begins at random throughout the thickness of the final sample at the same temperature, although the centre may lag about 20° C. with regard to complete recrystallisation. It thus appears that the sluggish recrystallisation of the central core is to some extent due to a residual effect from the hot-rolling process, and it can be reduced by the introduction of an intermediate anneal. If an intermediate anneal for a long period at 500° C. is applied, this legacy from hot-rolling can apparently be entirely removed, as in the final sample there was no difference between surface and centre with regard to the temperature for the start, or for the completion, of recrystallisation. Furthermore, the difference in temperature between the beginning and end of recrystallisation is generally smaller than with hard-rolled material or with material rolled after an intermediate anneal at 350° C. Samples intermediately annealed at 500° have, however, a much higher recrystallisation temperature than those intermediately annealed at 360° C. ; the difference is so great that recrystallisation is usually complete in the latter before it begins in the former, when heated simultaneously.

Thus, there are four outstanding features in the recrystallisation of rolled aluminium of commercial purity :—

(1) There is a pronounced difference in recrystallisation temperature between the centre and surface of hard-rolled strip.

(2) There is sometimes a minimum temperature in the recrystallisation/reduction curve at 75–80% reduction.

(3) The recrystallisation temperature is lowered by the introduction of an intermediate anneal at 360° C.

(4) The recrystallisation temperature is raised by the introduction of an intermediate anneal at 500° C.

Structure of Recrystallised Aluminium

When aluminium of commercial purity is completely annealed, the new crystals are, in general, oriented at random. In samples with a relatively high iron content, however, rolled without intermediate annealing, there is a tendency for a preferred orientation to exist which is

similar to the rolling orientation. This may be due to either incomplete recrystallisation or to the preferential growth of some of the new crystals with the same orientation as the worked material, though micro-examination suggests that recrystallisation is complete. The introduction of an intermediate anneal enables random orientation to be readily obtained during the final anneal, most probably owing to the small degree of fibring in intermediately annealed material.

Another exception to the general rule is that with high-purity aluminium (99·9%) a new type of preferred orientation sometimes develops, in which a cube face is parallel to the plane of rolling and a cube edge longitudinal. It may be supposed from the recrystallisation behaviour of the large crystal described earlier, that the preferred orientation developed during annealing may be due to preferential recrystallisation at the centres of old grains, so that it should predominate in metal liable to be coarse-grained at some intermediate stage—i.e., in relatively high-purity metal. With super-purity aluminium this orientation is encountered more frequently, and Fig. 15 reproduces X-ray transmission patterns taken from the same area of a specimen after heating to high temperatures. (The eight distinct maxima on the 311 Debye-Scherrer ring are characteristic of this orientation.) Some unfinished experiments suggest that the development of this orientation in super-purity aluminium may be avoided or suppressed by the use of suitable rolling and annealing conditions. In other words, factors other than a coarse grain at an intermediate stage may operate.

Films of Increased Protective Value on Elektron AM503

A Chromatic Bath of Reduced Acid Content Reduces Metal Loss*

OF the various protective treatments which have been devised for use with magnesium base alloys, says C. J. Bushrod,† none has achieved such general popularity as the acid-chromate bath. This bath, consisting of a solution of 150 grms. potassium dichromate in 1 litre of water acidified by the addition of about 200 cc. of nitric acid of sp. gr. 1·42, recommends itself on account of the following properties :—

(*a*) It forms good protective films on all magnesium base alloys.
(*b*) The conditions for correct film formation are not critical.
(*c*) The process is rapid and simple.
(*d*) The film provides a good base for paints.
(*e*) Satisfactory processing can be achieved with a minimum of preparatory cleaning.

The drawbacks associated with this process are twofold :—

(*a*) The cost of the solution, in relation to its service life, is somewhat higher than that of certain other satisfactory protection methods (e.g., the R.A.E. 30 mins. boiling bath).
(*b*) There is a small dimensional loss, of the order of 0·0005 in., from each surface during the treatment.

The commercial success of the acid-chromate treatment, over a number of years, is sufficient evidence that the advantages of the process far outweigh the drawbacks : nevertheless in certain cases the loss of metal during treatment of the parts is sufficient to warrant the use of other and less convenient methods. Thus where components are machined to fine tolerances before receiving any protective treatment this process is not applicable and, in this country, it is then usual to make use of the R.A.E. 30 mins. boiling bath, which does not cause any appreciable dimensional change.

The acid-chromate process has been of particular value in the protective treatment of magnesium alloy fuel and oil tanks for aircraft. These tanks are usually of all-welded construction and are made from rolled sheets of the binary magnesium-manganese alloy, Elektron AM503 ; the small dimensional changes accompanying the treatment are not of great importance on such components. During recent months, however, the position has been complicated by the development of fuel tanks. which are so large that the application of the chromating treatment is unavoidably prolonged owing to the length of time required to immerse the tank in and fill it with the solution. Thus in such cases certain parts of the structure are in contact with the acid liquor for times much longer than those required for film formation, and the dimensional changes occurring in these areas are correspondingly increased. It was thus apparent that, if these large fuel tanks were to be chromated as rapidly as possible in a cold bath, some modification of the treatment was necessary.

It seemed reasonable to assume that, if the acid content of the acid-chromate bath were reduced there would be a corresponding reduction in the dimensional changes which occur during the process ; if the films formed in such baths possessed good protective powers, and if in addition the reduced acid content did not unduly curtail the service life of the bath, then such solutions would be of special value for the treatment of bulky components.

Initial experiments were designed to find out whether :

(*a*) Films formed in baths of reduced acid content had protective values comparable with that of the standard film.
(*b*) The metal losses occurring during treatment of parts in these modified baths were sufficiently reduced to offset a possible reduction in the protective value of the films formed.

In carrying out these pilot tests four baths were used, each consisting of 150 grms. potassium dichromate dissolved in 1 litre of distilled water, to which was added 50 cc., 100 cc., 150 cc., and 200 cc. of nitric acid (sp. gr. 1·42) respectively. Test specimens of Elektron AM503 were cut from one piece of 20 s.w.g. annealed sheet, and the original chromate coat was removed by gentle abrasion with 1F emery paper. Twenty-four specimens were prepared, each specimen measuring 3 in. × 1·5 in. and having a hole $\frac{5}{16}$ in. diameter drilled in the centre of one short edge.

The results of the experimental work, indicates that acid-chromate baths containing about 75 cc. of nitric acid to each litre of water, will give the required protective film and have a life of length to warrant commercial application.

* British patent application 7233/42 (provisional).
† *Magnesium Review*, Vol. III, No. 3, 60–69.

The Iron and Steel Institute

Seventy-fourth Annual Meeting

THE annual meeting of the above Institute was held at the Offices of the Institute on May 13, 1943, Mr. James Henderson, president, occupying the chair. The president opened the meeting by extending a hearty welcome to all members present, and particularly to representatives from America, China, Luxembourg, Poland and other allied countries.

After the minutes of the previous meeting, held at Sheffield, had been taken as read and confirmed, the president reported, with regret, the death of a number of members of the Institute, including two honorary vice-presidents, Mr. Charles John Bagley and Mr. James A. Farrell. Mr. Bagley, who passed away on March 1, 1943 at the age of 95, was the senior member of the Institute by very many years. He had a distinguished career in the steel industry on the North-East Coast, serving with the Consett Iron Works and Tees Bridge, and later as a director of South Durham and managing director of Cargo Fleet. Mr. Farrell, who died on March 29, 1943, at the age of 80, had a distinguished career in America. In 1903 he became president of the United States Steel Products Export Company, and in 1911 president of the United States Steel Corporation, a position he held for 21 years. He was an honorary vice-president of the American Iron and Steel Institute. It was unanimously agreed to send a cable of condolence to that Institute.

Before presenting the report of council, Mr. Henderson conveyed congratulations to Dr. A. McCance and Dr. C. Sykes on their election to the Royal Society. He wished them many years of health and prosperity. The president also announced that the Council had decided to nominate Sir Charles Wright to be an honorary member of the Institute. In making this announcement reference was made to the unique distinction of Sir Charles in serving as Controller of the iron and steel industry during both wars. The Institute also is much indebted to him, said Mr. Henderson, for constant interest and assistance; It is just ten years since he retired from a most successful presidency, and it is fitting that we should add his name to the short and distinguished list of honorary members.

Membership of Council

The president announced the nomination of Mr. W. J. Brooke as honorary vice-president in recognition of his many years of distinguished service which he has given to the industry and to the Institute. Reference was also made to Mr. Richard Mather, who was elected a member of the Council last autumn, and to Mr. Desmond Lysaght, who was elected to the Council at the present meeting. Mr. Kilby and Mr. Glenn were also welcomed as honorary members of the Council in succession to Mr. Robinson and Dr. Gregory during their periods of office as presidents of the Lincolnshire Iron and Steel Institute and the Sheffield Metallurgical Association respectively.

Award of the Bessemer Gold Medal

The Council decided to award the Bessemer Gold Medal for 1943 to Mr. J. H. Whiteley. In announcing this award Mr. Henderson said: "Mr. Whiteley is the author of more than twenty excellent papers, and I must remind you that a term of the original conditions governing the award of this medal is that it must be presented for a paper read before the Institute and have special merit and importance in connection with iron and steel manufacture. It would be invidious to name a paper out of the many that he has contributed, but two aspects of his work to which I wish to refer in particular are his study of the acid hearth and slag and his examination of the nature of inclusions in steel. He received the Carnegie Gold Medal in 1917 for a paper entitled "The Eggartz Test for Combined Carbon in Steel," so that he started his career of distinction some time ago."

Unfortunately, Mr. Whiteley was not present at the meeting; he had another important appointment.

Williams Prize

No award of the Williams Prize was made in 1942, but the paper by Mr. R. L. Knight, of the Australian Iron and Steel Co., Ltd., presented at the present meeting, was received last year. and the Council decided to award the Williams Prize for 1942 to Mr. Knight. For the current year, 1943, the Williams Prize was awarded to Mr. A. Jackson, of the Appleby-Frodingham Steel Co., Ltd., for his paper on "The Linings of Large Basic Open-hearth Tilting Furnaces."

Presentation of Papers

Following the adoption of the report of Council and statement of accounts, two papers were presented for discussion; these were the "First Report of the Marine Corrosion Sub-Committee," which was presented on behalf of the Sub-Committee by Dr. G. D. Bengough, F.R.S., chairman, Dr. J. C. Hudson, and Dr. J. E. Harris; and "A Review of Basic Open-hearth Practice at an Australian Plant." In the absence of the author, this latter paper was presented by Mr. A. Robinson, chairman of the Open-Hearth Committee of the Iron and Steel Industrial Research Council.

In introducing the report of the Marine Corrosion Sub-Committee, Dr. Bengough pointed out that it should be regarded as a preliminary one; it gives a picture of the problem as it appears to the Sub-Committee and an account of the work done by the Sub-Committee to about the end of last September. The problem is very complicated and involves biological studies as well as physical. It was necessary to define the problem before it could be attacked. In this report no attempt is made to give solutions of the problem; it is too early to do so; the solutions—or preliminary solutions—will be given in the next report, a considerable amount of work for which has already been done.

Dr. J. C. Hudson introduced the section in the report which deals with raft tests. Since the report was prepared, certain test results can be made a little clearer by expressing them not in terms of "very good," "good," "moderate" and "bad," but in terms of the life to failure of the various specimens.

Dr. J. E. Harris introduced the anti-fouling section

of the report, and said that in the first place it was desirable to find out what the fouling organisms were. Out of a large number of organisms which settle on rocks and various other substrata in the sea, only a few ever foul ships. It was also desirable to find out the conditions of season of the year, light intensity, salinity and other factors which govern the settlement of these organisms. Then there was the important point of deciding whether there was any significance in the order in which the organisms settled on experimental panels. Dr. Harris summarised some of the work done, and indicated the lines on which work is proceeding.

Dr. W. H. Hatfield considered this subject of fouling as a problem of first importance from the standpoint of the war, and research involved considerable expenditure. Fortunately, this aspect has been adequately covered, and Dr. Hatfield was able to announce that this research is being conducted with no anxiety at all as regards the money required for the work.

Several members took part in the discussion, including Mr. J. Sinclair Kerr, Mr. F. C. Dycee-Teague, Mr. H. R. Wood, Mr. Thomas Jolly, Mr. Vernon Harboard, Mr. A. Robinson, and Mr. L. Ripley.

Mr. A. Robinson, in introducing the paper by Mr. Knight, congratulated the author, and discussed several aspects of open-hearth practice described in the paper. Detailed aspects were also discussed by Dr. E. Gregory Mr. L. A. Reed, Mr. Richard Mather, Dr. T. Swinden, Dr. J. H. Chesters, Dr. D. Binnie, Mr. T. Jolly, Mr. J. N. Kilby, and Mr. R. H. Myers. All were appreciative of the paper and of the spirit of enterprise and energy displayed at the plant described.

First Report of the Marine Corrosion Sub-Committee*

Of the Corrosion Committee†

The immediate scope of the above Sub-Committee's activities includes an intensive study of the forms, rates and mechanisms of fouling, together with extensive investigations directed towards the establishment of improved formulations both for anti-corrosive and for anti-fouling ship's bottom compositions. In addition to the introduction, which is presented in three parts, this Report details experimental work involving raft tests, and gives results in a section comprising nine parts; a further section of four parts deals with anti-fouling investigations. The summary of the Report is given here.

THE Corrosion Committee of the Iron and Steel Institute and the British Iron and Steel Federation has been engaged for a number of years in a systematic study of the corrosion of iron and steel, and of methods for its prevention. Corrosion arising as a result of immersion in sea-water has naturally come under its attention, and certain accounts of several investigations in this field have already been published. Additional impetus has been given to this branch of their work during the past two or three years by the desire for a fundamental research into the prevention of fouling.

The two main evils that beset all ordinary irons and steels of construction when immersed in sea-water are corrosion and fouling, and the primary object of the Sub-Committee's work is to establish the most efficient and economic cures. Although it is probable that the prevention of fouling is the major problem in the later stages of a ship's history, when the hull has become seasoned and has accumulated a thick coat of paint, the prevention of corrosion is of corresponding importance during its early life, both because serious damage may result from excessive attack and because the fouling problem cannot be solved if the hull is inadequately protected. Thus, the ideal solution is to devise methods of preventing both evils from the beginning.

Revolutionary methods for dealing with the problem should not be regarded as out of the question, but they are essentially matters for the future. For the time being the Sub-Committee have considered it wise to study and perfect the existing method. In essentials, this consists in applying a suitable protective coating to the steel plates to prevent them from rusting, over which, in turn, a toxic agent is applied to poison any fouling organisms that may try to settle on the surface. Thus, the immediate object of the Sub-Committee is to establish by experimental inquiry formulations and specifications for ships' bottom compositions, both protective and anti-fouling, the suitability of which, under given conditions, will have been demonstrated by experimental test. By "formulations" is understood not only precise details of chemical composition, but such further information concerning physical constants of medium, pigment and anti-fouling materials, etc., as may be required to enable any qualified manufacturer to produce the materials successfully.

The earliest experimental work of the Sub-Committee was undertaken in collaboration with the Admiralty Corrosion Committee, and consisted in comparative trials of various proprietary naval compositions. These trials took the form of raft tests on specimens of painted ships' plate exposed at Plymouth and Caernarvon. Raft tests, under the care of Dr. J. C. Hudson, are now playing an important part in the development and testing of paints of known composition as part of the Sub-Committee's systematic search for reliable compositions of stated formulation. The various series of raft tests are considered together, they form the subject of one important section in this comprehensive report. Another section is devoted to a description of the anti-fouling investigations proper, with particular reference to the biological line of attack, which is being conducted by Dr. J. E. Harris.

* Paper No. 9/1943 of the Corrosion Committee (submitted by the Marine Corrosion Sub-Committee) and presented at the Annual General Meeting of the Iron and Steel Institute, May 13, 1943.

† A Joint Committee of the Iron and Steel Institute and the British Iron and Steel Federation reporting to the Iron and Steel Industrial Research Council.

Raft Tests

Details of the experimental procedure adopted in the conduct of the raft tests are given and followed by a general description of the first two series of tests, which were made at Plymouth and Caernarvon on samples of proprietary painting schemes supplied by 23 different manufacturers. Some general observations are made concerning the formulation, media, pigments and toxins of these paints. Marked differences in consistency were observed from one sample to another, and the desirability of establishing a more uniform standard in this respect is stressed.

The performance of these comme cial painting systems varied considerably ; under raft-test conditions, when the rate of attack is accentuated as compared with service, the worst of them broke down within a few weeks when applied to new or even to repainted steel ; the durability of the best exceeded one year.

An analysis of the development of fouling during the Caernarvon test showed that all types of fouling—slime, weed, hydroid and shell—were observed, that there was a seasonal effect in this development, and this was also affected by the distribution of light and by subsidiary incidental effects due to tidal movement.

Comparison of the Plymouth and Caernarvon tests showed that the results were reasonably reproducible. The main " ranked " painting schemes at Plymouth were also ranked at Caernarvon. As expected, there was a general improvement in performance when the painting schemes were applied over old paint, as compared with freshly pickled steel. At Caernarvon the average lives to failure in these two cases were 16·2 and 10·7 weeks, respectively. Moreover, whereas the majority of failures on repainted surfaces were due to fouling, corrosion was the primary cause of failure on repickled ones. This emphasises that comparisons of anti-fouling compositions as such should be made only over satisfactory protective coats.

The life to failure by fouling, as determined at Caernarvon on repainted specimens, varied from a minimum of 5 to a maximum exceeding 56 weeks for the different proprietary compositions. There was a general but by no means exact correlation between these lives and the corresponding toxic indexes. (As described in the succeeding section, the toxic index is the weighted aggregate percentage of copper, mercury and arsenic in the anti-fouling composition, the mercury content being multiplied by 2 ; arsenic (as As_2O_3) is almost ineffective, but in these tests a weighting factor of $\frac{1}{3}$ has been assumed.) It is considered that several of the most marked deviations from this correlation are due to the presence of organic toxins in the composition. The test shows that on the average in order to ensure a life of six months under the experimental conditions at Caernarvon a toxic index of at least 30 would be needed. This would correspond to the presence in the anti-fouling composition of 15% of mercury as mercury compounds, or 30% of copper as copper compounds, or of an equivalent mixture, with the reservation that it is doubtful whether mercury can completely replace copper.

With two painting schemes, the drying time allowed between putting on the anti-fouling composition and immersion was varied from $\frac{3}{4}$ to $19\frac{1}{2}$ hours, as compared with the standard interval of 5 hours for the main test. These differences were not observed to have any systematic effect on the results obtained.

The variation in the protective properties of the painting schemes was as marked as that in the anti-fouling properties. The best were but slightly affected even after a year's immersion, whereas the worst suffered severe corrosion in only a few weeks. This variation is illustrated by resu'ts for the adhesion of the paints after test and the loss in weight of the various specimens on exposure. In terms of steel the latter varied from nil to 1·48 oz. per sq. ft. at Plymouth, and from 0·02 to 1·49 oz. per sq. ft. at Caernarvon ; the durations of exposure were 29 and 26 weeks, respectively. Comparable losses for bare pickled steel exposed simultaneously were 1·78 and 1·39 oz. per sq. ft., from which it is obvious that the worst compositions were practically useless. In general, the correlation of the Plymouth and Caernarvon results for the painting schemes common to both series of tests was good—i.e., the orders of merit in regard to protective properties were substantially the same.

Another property investigated was the tendency for the painting schemes to promote pitting or local attack at " holidays "—i.e., gaps in the coating when the steel was left bare. This was done by leaving small areas, $\frac{1}{2}$ in. square, unpainted when applying the coatings. It was found that the depth of pitting at these holidays after six months' exposure ranged from 0·007 to 0·080 in. Moreover, the results were substantially the same in both test series, from which it is concluded that the degree to which pitting is promoted at bare areas in a paint film is a characteristic property of the painting scheme used. At Caernarvon the worst pitting was associated with a scheme in which two coats of red-lead / oil paint were used as the protective immediately beneath the anti-fouling composition. This does not necessarily mean that red-lead primers should not be used on ships' hulls, but it does point to the desirability that they should be suitably formulated for this purpose.

In general, it is concluded from these two series of experiments that, although service tests must remain the final criterion of performance, particularly where anti-fouling properties are concerned, raft tests should give a good indication of the relative merits of protective compositions, and the results of these initial series should bear a substantial relationship to the probable behaviour in the case of painting schemes applied to new construction. New series of tests will include a study of the influence of the build-up of the coating resulting from successive repaintings.

Subsequent series of tests have been conducted on more fundamental lines. Series IV, begun at Caernarvon in January, 1942, is designed to elucidate the effect of various methods of surface preparation, weathering, pickling, etc., coupled in some cases with temporary protective treatments applied during the weathering period (corresponding to exposure on the stocks) before the painting proper. Data are given for the loss in weight and, in the case of as-rolled plates, for the percentage descaling of the steel as a result of this preliminary exposure. These show that for the general run of ships' plates at least six months, and probably much longer, would be needed to remove the mill scale entirely by weathering ; further, the weathering process is accompanied by the formation of substantial quantities of rust which cannot be removed completely from the surface by hand-cleaning.

All the temporary protective treatments tested, with the exception of mineral oil, were effective in eliminating rusting almost completely during the weathering period.

Mineral oil had practically no protective effect and was markedly inferior to boiled linseed oil, which, when rubbed on thinly at the rate of about 3,000 sq. ft. per gal., was found to protect the plates adequately during 70 days' exposure in the open at Caernarvon. Protection, however, is not the only criterion ; it is equally important that the temporary protective shall be compatible with the final painting scheme. Interim results for the immersion tests proper show that the behaviour of one of the schemes used was adversely affected both by the boiled linseed oil film and by a temporary protective coat of red-iron oxide paint (in an ordinary oil medium) ; on the other hand, the application of a temporary protective coat of thin red-lead paint, which dried out thoroughly during the weathering period, resulted in a pronounced improvement in performance. This again points to the necessity for complete control of the formulation of all paints used for marine purposes.

In the case of a moderately good painting scheme, better results were obtained both on as-rolled and on pickled surfaces that had been weathered for 105 days than on a freshly pickled surface, possibly because the presence of rust improved the adhesion of the paint. This may not be a general result, because with a better painting scheme which shows little deterioration to-date the reverse is the case, the freshly pickled specimens being the best of the batch.

Examination of specimens removed after 24 weeks' immersion showed that in all cases where the final scheme had been applied over surfaces from which mill scale had been incompletely removed by weathering, pitting had occurred at the boundaries of the remaining mill scale up to a depth of 0·34 in. Pickled surfaces suffered a different type of attack, which was localised to a shallow depth over comparatively large cellular areas, corresponding with places where the paint had flaked off. Since the tendency to pit will remain so long as the mill scale is incompletely removed, and since weathering in the yard and on the stocks cannot be relied on to bring about complete descaling, it is suggested that more attention be devoted to the use of pickling for this purpose.

Good agreement in assessing the protective values of the various painting schemes was observed between the results of the raft tests and those obtained in laboratory tests at the Chemical Research Laboratory, using a special apparatus in which specimens are rotated rapidly in sea-water.

Series V, also begun in January, 1942 is concerned with new types of protective paints that are being developed by Dr. U. R. Evans. These are designed to overcome one of the fundamental disabilities associated with the protective painting of structural steel, namely, that all coats should be applied to a good surface under good conditions. It is hoped that these new paints will prove insensitive to these adverse factors, and that they will be found to give good results when applied, for instance, to wet and rusty steel. Results so far are encouraging, and in tests of up to six months the protective properties of some of them have proved equal to those of the best commercial compositions.

Tests on a number of formulated compositions prepared by the Paint Research Station, through the courtesy of Dr. L. A. Jordan, are reported (Series VI). These constitute an investigation of the possibility of substituting coumarone and lanolin for rosin and linseed oil, respectively, in media of the rosin/coal-tar putch/linseed-stand-oil type. These schemes failed rapidly by fouling, largely because the toxin contents of the anti-fouling compositions were too low, but the results showed that several of the schemes in the substitute media had promising protective properties.

This brings the account of the raft tests up to date. An outline is given of the immediate experimental programme, which will consist of three interrelated series of tests on special compositions formulated and prepared by the Sub-Committee :—

(1) Tests at Caernarvon, in which the pigmentation of the protective composition will be the main variable. This series has just been begun and a few details of it are given.

(2) Tests at Emsworth, where a new testing station will be set up by the Admiralty Corrosion Committee under the direction of Dr. I. G. Slater. Here the main variable will be the medium used for the protective compositions ; a limited number of alternative pigments will be put up in each medium. In addition, the effect of building up thick films of paint on the steel specimens will be systematically investigated.

In series (1) and (2) the same anti-fouling composition will be used throughout each series.

(3) Tests at Millport, by Dr. J. E. Harris, on anti-fouling compositions, in which a number of old and of possible new toxins will be tested in a variety of media.

All these series will be correlated with each other and with the results of accelerated laboratory tests at the Chemical Research Laboratory, Teddington, by means of suitable controls. In addition, storage tests, both at ordinary and at tropical temperatures, will be made to test the keeping properties of all the compositions prepared.

Section B concludes with an account of subsidiary series of tests relating to protective metallic coatings and to unprotected irons and steels, respectively.

Immersion tests at Gosport have shown that thin coatings of zinc—or aluminium—ranging up to 0·005 in. thick, protect steel adequately for at least one year (the tests are still in progress). Zinc coatings have a decidedly toxic effect as compared with aluminium, but this does not suffice to prevent fouling completely. Tests at Caernarvon in which paints were applied over the metallic coatings before immersion showed that the effect of zinc on fouling was evident even when the surface was painted over with three coats of naval compositions, finishing with an anti-fouling coat. The behaviour of this painting scheme was better over the metallic coatings than over bare steel, no doubt a result of the suppression of rusting ; there was evidence, too, of better behaviour on a surface pickled by the Footner process, which includes a finishing dip in dilute phosphoric acid, than on a surface pickled by the normal Admiralty procedure in cold dilute hydrochloric acid.

Tests on bare irons and steels at Gosport and at Plymouth showed that, apart from the fact that Scottish wrought iron was about 15% more resistant to corrosion than ordinary mild steel, there was no appreciable difference in the losses in weight of numerous types of ordinary and low-alloy irons and steels of constructional quality when immersed in sea-water in the bare condition. In particular, the presence of small amounts of copper, alone or in combination with small amounts of chromium, had no effect. This should be regarded as

an interim result, but it is clear that any differences will be negligible when, as is imposed by necessity in service, the irons or steels are exposed with a protective coating.

Anti-Fouling Investigations

The biological aspects of fouling are first considered. The community of organisms appearing on non-toxic surfaces includes marine bacteria, seaweeds, diatoms and sessile marine animals, such as barnacles, calcareous tubeworms, ascidians (sea-squirts), polyzoa, mussels and hydroids. The community can be resolved into fairly well-defined separate layers; from a basal carpet layer a more complex community may grow up. The extent of this growth varies with the toxicity of the paint, with the season of immersion and with the organisms available to settle and able to maintain themselves in a community. The presence of some types of fouling impedes the settlement and growth of other types.

Different types of organisms differ widely in their sensitivity to the metallic toxins commonly used in anti-fouling compositions. This is demonstrated by the results, which have also been used to study the effects of different toxin concentrations, differences in medium and other variables on the performance of specially formulated anti-fouling compositions.

The development of fouling, i.e., the sequence in which the various types of organism appear on the surface and their relation to each other, is being studied. The earliest stage consists of a bacterial or diatom slime. This film has been observed to form on all anti-fouling paints exposed in the sea. The amount of slime bears a definite relationship to the toxicity of the paint, and is also influenced by some of the ingredients in the paint medium. A quantitative method of determining slime formation has been developed, in which the film is stained with methylene blue; the dyestuff absorbed is leached out and estimated colorimetrically. It is hoped to make use of this as a rapid means of determining toxicity, since the time necessary for the formation of slime is a few days only and much shorter than that required for the development of weed-fouling. Moreover, slime formation is much less affected by the season, which should render it possible to conduct determinations all the year round without reference to the fouling season.

The diatom flora also provides a means of estimating the toxicity of a surface soon after its immersion in sea-water, long before visible signs of fouling on a large-scale appear; this method, too, should be capable of use at all times of the year.

Evidence shows that the toxicity of a surface has little effect on the growth of an organism once attachment has taken place. Possibly attachment could be prevented by the incorporation in the anti-fouling composition either of substances that repel or kill the organisms in their settling stages, or, alternatively, of substances which prevent the adhesion or setting of the cementing substance by means of which the organisms attach themselves. The second possibility is being explored with special reference to the cement secreted by the larva of the common barnacle, one of the most important organisms. This study has been greatly facilitated by the development of methods for rearing this animal in the laboratory from its free-swimming larval stage to its fixed stage.

Fundamental investigations of the principles underlying the action of anti-fouling compositions are in progress. These consist mainly of raft tests on formulated compositions, a large number of which, containing both inorganic and organic toxins, have been prepared.

In the case of the common inorganic poisons, it has been demonstrated that mercury is approximately twice as effective against weed fouling as copper (weight for weight); arsenic, at least in the form of arsenious oxide, is almost ineffective. The ratio of the effectiveness of mercury and copper is higher in the case of barnacle fouling, for which it is at least three to one; again, arsenic has little effect.

The anti-fouling effect of the usual copper compounds was primarily determined by their copper content and fell off in the following order: Copper bronze, cuprous oxide, copper thiocyanate, Paris green. The physical form and solubility of any given toxic pigment will also be of importance.

Numerous organic toxins have been tested, both in the laboratory and in the form of paint films on panels immersed in the sea. The laboratory tests consist in determinations of their toxic effects on marine organisms in sea-water cultures; under these conditions many of the compounds are at least as toxic as copper, and some even more toxic than mercury. On the other hand, with a few promising exceptions, the toxicity of these compounds when incorporated in actual paint films has fallen far below the values that would be expected from the sea-water culture experiments.

Chemical studies of the rate of leaching of copper, mercury and organic compounds from paint films in sea-water are in progress, and micro-analytical methods have been developed for this purpose.

The maximum effectiveness of the most suitable toxins, when they are discovered, will only be achieved by their incorporation in the most appropriate paint medium. This aspect of the matter is being studied, due regard being paid to the conflicting requirements that the toxins shall be readily and continuously available whilst the coating itself shall be as durable as possible.

Attempts are being made to ascertain the chemical mechanism by which toxins produce their anti-fouling action. To this end the chemical equilibrium between the copper pigment, the free fatty acid in the paint medium and the sea-water is being investigated.

The importance of providing efficient protective undercoats is stressed. Statistical analysis of the correlation of fouling with rusting on a number of specimens coated with proprietary painting schemes showed that fouling always followed rusting. No evidence was obtained that the presence of fouling organisms on the specimens induced rusting.

Section C concludes with some notes on the results of the examination of upwards of 60 samples of fouling from ships. The number of species represented was fewer than is found on specimens exposed in raft tests probably because of the more limited flora available for settlement in many ports compared with those in the vicinity of the rafts. Only certain seaweeds have spreading attachment systems and growth forms suitable to withstand the force of water movement over a ship's hull. Considerable progress might be made by concentrating attention on the elimination of the specific types of organisms chiefly responsible for fouling under practical conditions. This, in turn, suggests that anti-fouling formulations of the future will not be one but several, each specially designed for the special conditions of service of individual vessels.

A Review of Basic Open-Hearth Steel Practice at an Australian Plant*

By R. L. Knight

In this part of the author's valuable paper, presented at the recent Annual General Meeting of the Iron and Steel Institute, some practical aspects are given of basic open-hearth operations at the steelworks of Australian Iron and Steel Limited. Attention is directed to refractories; furnace performance; operations, including charging, melting, finishing the heat, and tapping; pit practice; ingot moulds; mould preparation, pouring practice, and maintenance.

Refractories.

AUSTRALIAN steelmakers, in common with those of Britain and the United States, are giving much attention to open-hearth refractories. Possibly the foremost future saving in costs will be derived from the use of improved refractories. There can be little doubt that the lowest open-hearth operating costs will not be realised until the all-basic furnace appears. To-day, five of Port Kembla's seven furnaces are all-basic except for the roofs and downtake arches. Campaigns close down when the roof structures become unstable and dangerous. When silica roofs are replaced by basic arches constructed with proven high-strength thermally satisfactory basic brick, campaigns of 400 heats may lengthen to possibly 1,000. The balance of present design will then be upset, and weaknesses will show at other sections of the structure. It is believed that some of the Kembla units could use basic roofs to advantage now. A full campaign would run up to two years, divided into two non-stop periods with a short cleaning and patching time between. Steelmaking rates would be 10—15% faster, and overall repairs and maintenance costs appreciably lower.

Silica brick of the following composition is used in roof construction :—

SiO_2.	Fe_2O_3.	Al_2O_3.	TiO_2.	CaO.	Specific Gravity.	Fusion Point.	Porosity.
95%	1%	0·6%	0·1%	3%	2·34	1,740° C.	22%

The physical properties of the brick are well up to accepted standards, but not equal to those of the "super-silica" brick available in some places abroad.

Roof life varies with the type of fuel fired, and on the present basis of comparison averages 370 heats on producer-gas-fired units. The traditional method of recording roof performance by the number of heats carried is wrong. Some plants during a campaign entirely replace a roof in running and mid-campaign repairs, and credit the service to one roof. This procedure, of course, is misleading, since the roof cost must greatly exceed the expenditure on the original roof. Further, roof thicknesses vary from plant to plant, and it is obvious that a run of 300 heats with a 16-in. roof at one plant is a better performance than a run of the same number of heats with a 22-in. roof at another plant. Steelmakers quoting furnace performance should therefore refer to roof life in terms of weight of silica refractory consumed per ton of ingot.

Front Walls, Back Walls and Bulkheads.—For some years past a highly efficient domestic refractory has given good results in the construction of front walls, back walls, air and gas bulkheads, cold walls and air-cooled monkey walls. Australia has not progressed as rapidly as the United States and Britain in the development of basic refractories and supplies of local basic bricks of specified quality are not available. However, chrome-magnesite pipes made at the open-hearth have run for 800 heats in the air bulkheads of a 230-ton tar-fired furnace without repair. This wall was 18 in. thick. Old boiler tube of 1—4 in. dia. is oxy-cut to the desired length and filled with 50% chromite, $\frac{1}{4}$ in. maximum, and 50% dead-burned magnesite, $\frac{1}{8}$ in. maximum. The bond is a 4% solution of 1·7 sp. gr. waterglass, and the materials are mixed in a concrete mixer. The packed pipes are air-dried for at least three months, and finally dried in a skip on a furnace roof before use. Masonry built of pipe is bonded with the same mixture. Rarely before 150 heats is hot repair to a front wall or straight back wall required, and up to 200 heats have been made before patching. The average front-wall life is 175 heats.

Midfeather Walls.—Recently an all-basic-end gas-furnace was put into service with midfeathers constructed of hard-burned chrome-magnesite brick. After 16 weeks' operation these walls are still perfectly gas-tight, and it is hoped that midfeather leaks at last have been overcome. From experience at Kembla with basic ends, it is not difficult to anticipate an early release from the exasperating task of removing those big masses of pocket slag that too often have kept furnaces off the line for days beyond the normal repair periods. These dense slags have been replaced by soft easily removable accumulations on tar units; it is hoped that a similar result will be obtained in the gas-fired furnaces.

Regenerators and Slag Pockets.—The slag-pocket and fantail arches are of silica brick construction, but regenerator arches are of standard fireclay bricks containing :—

SiO_2.	Fe_2O_3.	Al_2O_3.	TiO_2.	CaO.	MgO.
71%	2%·8	21%	1%	0·4%	0·5%

Checker tiles of a similar material are of two standard sizes, 15 in. × 6 in. × 2 in. and 13 in. × 6 in. × 2 in., laid in uniform box-flues, one pass only. This plant has not experimented with tiles of variegated design or multiple-pass checkers. Careful attention to cleaning, with trained crews working to established weekly schedules, and the education of first helpers to prevent overheating has ensured comparative freedom from checker troubles. Checkers are cleaned down with

* Continued from page 239 in the April issue.

compressed air, at 100 lb. pressure per square inch, through slotted cast-iron apertures along and across. Furnace crews can examine checkers with an observation pipe extending through the charging floor into the dome of the regenerating arch.

Furnace Performance.

The long-term performance of the modern Kembla furnaces has been encouraging. The details of the first campaign of F furnace, completed in January, 1940, are :—

Days operated	246
Heats	400
Tons per heat	203
Gross tons	81,362
Tonnage per week	2,315
B.Th.U. per ton of ingots	3,658,500

CHARGE COMPOSITION.

	Tons.	%
Mixed and direct metal	62,208	66·8
Cold iron	5,717	6·1
Steel scrap	16,462	17·7
Iron ore (60% Fe)	8,656	9·4
Limestone	3,343	—

The record monthly production during this campaign was made in March, 1940, when the furnace produced 11,311 tons of ingots at a rate of 2,862,000 B.Th.U. per ton.

The initial campaign of the latest Kembla furnace, G, is still in progress at the time of writing, but the mid-November, 1941, details were :—

Days operated	150
Heats	222
Tons per heat	233
Gross tons	51,789
Tonnage per week	2,415
B.Th.U. per ton of ingots	3,672,000

For July, 1941, G furnace made 11,627 gross tons of ingots at a rate of 3,442,000 B.Th.U. per ton, charging 15% of steel scrap.

Operations

Stockyard.—The stockyard lies parallel to the furnace building. A standard-gauge railway track runs through the yard and there are shunting spurs at each end. With the exception of ore, all charge components are marshalled within the yard by two 15-ton combination magnet and cradle cranes. Since the plant normally operates on a 15% scrap burden all scrap is mill return. Scrap returns are organised so that about 90% arrives at the yard ready for charging. Rolling stock for transport comprises light, flat-topped wagons which carry one charging crate with three pans. Each fabricated charging pan has a capacity of 36 cub. ft. and holds 1—3 tons of scrap, or 2½ tons of ore or 1½ tons of limestone.

Provision is made in the scrap-yard for the isolation of alloy scrap, in a coloured area.

When preparing charges the cranes usually work in collaboration, one loading with the magnet and the other elevating the loaded crates to the two 25-ton weighbridges at charging-floor level. After weighing, the crates are set down on the charging bench opposite the furnace.

A series of large bins is provided for phosphate rock, sea-sand, bauxite and fluorspar, and these are hand-loaded into pans as required.

As in all plants, the scrap is proportioned according to type, so that furnaces receive a balanced charge of light and heavy scrap.

Bottom Maintenance.—The quality, quantity and yield of steel made in any open-hearth shop is largely dependent upon bank and bottom maintenance. Loose, dirty bottoms tap metallurgically dirty steels, and bottoms which drain badly can lower the overall plant yield by more than 1%. Deoxidation practices are also directly influenced. The prevalent custom of allowing cost and availability to govern the choice of materials used for repairs is not commendable when it is possible for the cost of extra time lost by bank and bottom repairs through using cheap refractories to exceed the extra cost of better materials.

Severe bottom damage is repaired by sintering-in dry calcined magnesite with 10—15% of basic slag, a proportion of burnt dolomite occasionally being added to save time. Localised less severe damage is often repaired with wet calcined magnesite applied as a partially hydrated pug, and charging can commence almost at once without danger. Severe damage along the slag line is repaired with burnt dolomite, but routine fettling is done with crushed raw dolomite. When applied it must be calcined and sintered into place before the hot-metal addition. If fixation does not occur, the dolomite lifts and interferes with the flush slag. That portion which remains in the furnace increases the finishing slag volume, makes slag control difficult, and slows down production ; this perverse condition existing from heat to heat. At times 6 in. or more of the bank comes away. Moreover, the furnace cannot remain idle for any length of time without serious deterioration of the banks. The best fettling refractory would appear to be a self-bonding weatherproof ferruginous dolomite material in uniformly sized circular particles of high softening point.

The use of the fettling machine at Kembla undoubtedly saves time. Generally, bottoms tend to become high and banks to toe out, and periodical wash-outs are essential. The oldest bottom here is still giving satisfactory service after eleven years.

Charging.—The following is the charging time objective on a standard heat for a 230-ton fixed-fuel furnace :—

Begin charging	0 hr.
Scrap	35 tons.
Manganiferous ore	46 tons.
Limestone	11 tons.
Finish charging solids at	1 hr. 30 mins.
1st ladle hot metal	3 hr. 45 mins.
2nd ladle hot metal	4 hr. 15 mins.
3rd ladle hot metal	4 hr. 45 mins.
Total charging time	4¾ hours

Fig. 6 shows a portion of an actual time study of charging and filling-up times. Correlated with charge composition and hourly production rates, such graphs, recorded daily, are invaluable for the rapid analysis of shop operations.

As a general rule, one-third of the scrap is spread over the bottom, followed by the limestone, then the ore, and finally the balance of the scrap. The timing of the hot-metal addition is of great importance. If the furnace is filled too soon the heat is delayed,, insufficient flush slag is taken and the finishing slag is too siliceous. Should filling be delayed, violent reactions ensue and floor spills occur, yield is lost and the carbon contents of the heats at melting are irregular. For satisfactory routine operation and consistent quality, furnaces must take hot metal on time. The procedure at Kembla is to pour the iron from the 1,200-ton mixer to the transfer ladle, run the car to the weighbridge, 40 ft. away, and hoist

directly to the floor with the 100-ton crane. The following is a typical time cycle for hot-metal additions :—

Pouring from mixer, weighing, hoisting, transferring	14 mins.
Pouring into furnace	5 mins.
Returning ladle to mixer, lowering to bridge, and re-taring..	8 mins.
Total	27 mins.

The average scrap charge is 13—16% of the total metallics, and consists almost entirely of mill returns. On special grades of alloy steels heats are charged with up to 50% of scrap, mainly with a view to alloy recovery and controlling expensive alloy additions within narrow limits.

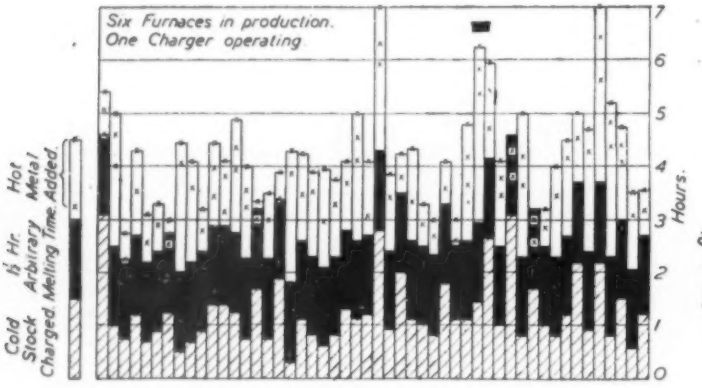


Fig. 6.—Charging time for fifty consecutive heats.

On the rare occasions when cold iron is charged it is placed on top of the scrap.

In order to obtain smooth operation, accurate melts and consistent quality, it is imperative that the proportion of silicon in the hot metal should be known to the melters beforehand. In addition to a calculated figure obtained from the mixer throughout, two mixer-iron analyses are made on each shift. The melter adjusts the order burden accordingly. The desired basic iron composition is as follows :—

Silicon.	Sulphur.	Phosphorus.	Manganese.
0·8—1·0%	0·04% max.	0·10% max.	2·3—3·0%

The average charging temperature of the mixer metal is 2,300° F. Fig. 7 shows the range of variation of silicon content of the mixer metal compared with the sequence of silicon contents in the iron from the blast-furnaces.

Melting.—Charges are melted down as rapidly as possible and arranged to melt at 0·4—0·6% higher than the final carbon desired, so that 1—1½% of manganiferous iron ore can be used. The average fuel input per hour varies between 55×10^6 and 65×10^6 B.Th.U., depending upon conditions. Heats are kept hot throughout, and the temperature for tapping is adjusted at the finish. The direction of fuel flow is reversed every half-hour early in the heat and at 15-mins. intervals in the later stages. Except in the case of ingot iron, it is generally plant practice to clear the bath thoroughly before attempting to ore. Over-oreing is avoided, and with the normal limestone burdens carried little flux is necessary to shape up the slags. On special and alloy heats fluorspar, sand and bauxite are used, with lump ore to break up the lime and incite the bath to action.

The rod and pour tests are used to judge tapping temperatures, and the melters check these with optical-pyrometric readings taken by the metallurgical observer.

The oxygen carried in the downtake gases varies between 1% and 4%, and each furnace is checked by Orsat tests at regular intervals to control combustion.

Finishing the Heat.—For testing the steel-bath samples there is a well-equipped laboratory on the outer end of the charging platform between E and F furnaces. Carbon is determined by colour and combustion methods

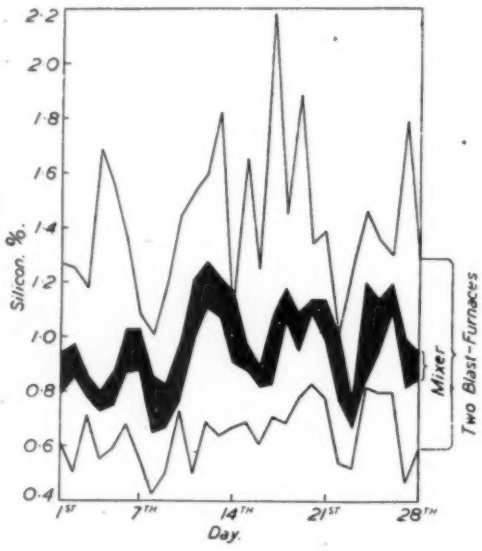


Fig. 7.—Daily range of silicon contents of blast-furnace and mixer casts.

by shifts of experienced chemists. Rapid accurate estimations are made of manganese, phosphorus, sulphur, and, when required, of nickel, chromium, etc. A slag-tester keeps the melter informed of the composition of the slag. Established rules concerning proximity of ore, additions to alloying or tapping are obeyed, and on all special heats the last few points of carbon are rodded out. Colour carbon results from carefully prepared samples are obtained in 15 mins., and combustion carbon figures 30 mins. after the samples are taken.

Tapping.—Present tap-hole practice has given good results for several years. The hole is stopped from the front with a mixture of 10% ground coal and 90% crushed burnt magnesite mixed to a stiff pug. From the pit-side the hole is filled with ½-in. burnt dolomite, and finally plugged with a ball of plastic clay. When opening the hole the ball of clay is removed, the hot dolomite raked out, and the chill opened with the oxygen lance. A pneumatic-tyred hand buggy, fabricated from buttweld pipe, carrying two bottles of oxygen, is used to run the gas to backstandings for use on the tapping hole. The diameter of the tapping hole is such as to allow the metal to run out at the average rate of 15 tons per minute. Tap-holes are re-made in the traditional way with a 4-in. or 6-in. pipe, depending on the speed of run-out desired for special heats.

Pit Practice

Steel from these furnaces is tapped into conical ladles of 160 tons capacity, 13 ft. 9 in. high, 11 ft. 7 in. dia. at the top and 10 ft. in dia. at the bottom. All ladles are double-stopper rigged. Ladle-linings are not insulated, but consist of a 3-in. safety lining of firebrick and a 6-in. working lining of special clinker common brick. Daubing of linings is not practised. The average number of heats per ladle lining is 12. Ladles are heated to 600° F. by coke-oven gas.

Stopper-rod assemblies are made up by one man. The rod, 15 ft. long and 2 in. in dia., takes a firebrick sleeve with a 2⅛-in. bore. Imported graphite stopper-heads are keyed on in the usual way, and the assemblies are thoroughly dried in a vertical oven fired with coke-oven gas.

Overflow and flush slag is taken off in 300-cub. ft. cast-iron thimbles and removed from the pit on Pollock-type cars. The pit is provided with the necessary accessories for the rapid cleaning up of scrap and debris.

Ingot Moulds

All the steel is top-poured. Table VI incorporates the details of mould design. All moulds have corrugated contours. Opinions differ on the relative merits of straight-sided and contoured ingots for blooming mill feed. On the other hand, multi-sided ingots are universally used for forging. Exact valuation of the advantages of a contoured milling ingot is different, but it is accepted here that with steels of similar quality, poured under precise conditions of temperature and speed, the corrugated ingot has a stronger skin and cracks less in the blooming mill than the straight-sided ingot. Corrugation adds nothing to the cost of the mould, and, provided that the design and mould-life are satisfactory, contouring is good insurance.

Numerous experiments have been made to determine the most satisfactory iron for mould manufacture. These have indicated that the composition of the iron can vary considerably without adverse effect. Moulds are cast with a direct-metal mixture of two parts of basic iron with one part of foundry iron, with selective provision that the manganese in the casting will not exceed the silicon by more than 1·2%. When both blast furnaces are producing high-manganese basic iron, cupola metal is added to obtain the desired composition. The following examples are typical :—

No. of Heats.	Silicon, %	Sulphur, %	Phosphorus, %	Manganese, %	Carbon, %
123	1·2	0·03	0·21	2·1	4·2
121	0·9	0·04	0·10	1·5	4·2
117	1·9	0·03	0·48	1·6	4·0
109	1·4	0·02	0·54	1·6	4·1
107	1·5	0·04	0·26	1·9	4·2
100	2·0	0·02	0·50	1·6	3·9

Tests on this plant have indicated that the rest period has the most pronounced influence on mould life. The period at Kemla is not standard, and may vary from 2—to 10 hours. Ingots should be stripped as soon as they may be, and the moulds allowed to cool to 150–200° F. before being used again.

Hot-top practice is varied. Alloy and some other special grades are poured into wide-end-up corrugated moulds seated on a spigotted stool. Superimposed fixed heads are used. Both ends of the mould and the base of the head casting are machined. Dead-killed spring steels are cast into wide-end-down moulds with a recessed tile head. This is a utility mould suitable for certain types of steel.

Mould Preparation

Perhaps the most difficult part of quality steelmaking is the pouring of consistent high-grade ingots with good surface characteristics. Surface quality is influenced by several factors, of which mould conditioning is one of the most important. Undoubtedly steel reconditioning costs vary directly with the care given to moulds and stools prior to casting.

Primarily, a high standard of cleanliness is essential. Secondly, mould wash, if used, should be carefully applied. Experience at Kembla has shown that no mould wash is better than an indiscriminate application. The choice of one from the numerous mould coatings available may depend upon the type of steel to be poured, and the method of pouring.

There is no standard mould coating at Kembla. Graphite wash with brine solution was tried and rejected. Dry tar spraying on at the right temperature does a good job. The coating should be thin, hard, and glossy. The best ingots here are cast in uncoated moulds. While hot from the previous heat the moulds are washed down with fresh water, which loosens accretions. When dry they are scrubbed down with long-handled wire brushes. Finally, compressed air is used to clean the mould

TABLE VI.—INGOT MOULD DATA.

Mould : / Used for :	25 in. × 21 in. Standard Wide-End-Down. Rimming and Semi Killed Steels.	29 in. × 24 in. Wide-End-Down. Large Beams.	38 in. × 20 in. Wide-End-Down. Standard Slab Mould.	25 in. × 21 in. Wide-End-Up. Killed Alloy and Special Steels.	25 in. × 21 in. Wide-End-Down. Spring Steel.
Walls	Corrugated.	Corrugated.	Corrugated. wide side.	Corrugated.	Corrugated.
Height	77 in.	80 in.	78 in.	—	77 in.
Overall mould section :					
Top	34½ in. × 30½ in.	42 in. × 36¼ in.	47⅜ in. × 31½ in.	—	36½ in. × 32½ in.
Bottom	39⅜ in. × 35⅜ in.	45⅝ in. × 40 in.	54½ in. × 36 in.	—	39⅜ in. × 35⅜ in.
Contour :					
Top	23⅜ in. × 19⅜ in.	28 in. × 22¼ in.	36⅜ in. × 18½ in.	—	Square tile head
Bottom	25⅛ in. × 21⅛ in.	29⅜ in. × 24 in.	38 in. × 20 in.	—	25⅛ in. × 21⅛ in.
Wall thickness :					
Top	5⅜ in.	7 in.	6½ in.	—	4⅜ in.
Bottom	7⅛ in.	8 in.	8 in.	—	7⅜ in.
Thickness of corners :					
Top	4⅛ in.	5⅜ in.	5⅛ in.	—	4⅛ in.
Bottom	5⅜ in.	6⅜ in.	7⅛ in.	—	5⅜ in.
Radii of corners :					
Top	1½ in.	2½ in.	2½ in.	—	—
Bottom	1⅝ in.	2⅝ in.	2⅝ in.	—	1⅝ in.
Taper per ft.	0·31 in.	0·263 in.	0·19 in.	—	—
Height, base to hot top	—	—	—	—	62 in.
Mould weight	5·05 tons	7·74 tons	8·6 tons	—	5·13 tons
Ingot weight	4·40 tons	5·79 tons	6·53 tons	—	4·37 tons
Weight ratio	1·15 to 1	1·335 to 1	1·225 to 1	—	1·17 to 1
Number of corrugations :					
Wide side	2	5	—	—	—
Short side	2	4	—	—	—
Ratded	Not	13½ in. × 1½ in.	Not	—	Not
Average mould life	97 heats	93 heats	95 heats	—	—

wall and the stool thoroughly. A few minutes before casting, a length of corrugated kraft-paper in the form of a large cylinder is dropped on to the stool. This has given good results in preventing stool splash. Stools are dished and spigotted to form an effective seal, and the small crevice between the mould and stool is filled with grains of ferro-silicon poured down a funnel.

Pouring Practice

Following the establishment of a satisfactory pouring practice, constant supervision is necessary to maintain a high standard. Consignments of nozzles, stopper-heads, sleeves, etc., should be carefully checked for quality and size. The workmanship in rod assembles must be inspected, stopper riggings on ladles kept in perfect condition, stopper and nozzle settings examined, and pouring crews encouraged to teem cleanly. To achieve a consistent standard entails a lot of persistent work. Nozzle sizes at this plant have varied during the past ten years, and from the experience gained precedure has been formulated.

All grades of steel are top-poured. For any grade there appears to be an ideal pouring temperature for a given pouring speed. Ingots so poured into scrupulously clean moulds of modern design and then satisfactorily soaked will roll with a minimum of cracking in the blooming mill. Should the heat be poured too hot or too fast, the results will be indifferent. For some time 1½-in. composite magnesite ring nozzles were used for most grades, with pouring speeds of 2·5—2·75 tons per min. Experiments proved that increasing the pouring rate to 4 tons per min. and lowering the pouring temperature enhanced the surface condition of the ingots without impairment of rolling quality. At present the standard shop nozzle is 1¾-in. plain fireclay, but the size ranges from 1½ in. to 2¼ in., according to the grade of steel that is to be poured. Rimming steel of 0·08% to 0·10% carbon is poured through 1¾-in. nozzles and at a temperature 40°—50° F. hotter than semi-killed steel of the same carbon content. This grade is allowed to rim-in until a central molten area of 9 sq. in. remains, when it is capped with rolled steel plate-caps weighing 150 lb.

Alloy steel also is teemed with 1¾-in. nozzles with careful control of the temperature. Transformer steel, containing 4·5% of silicon, is teemed through 2¼-in. nozzles after reladling twice through 3-in. nozzles, for cool steel and clean ingot surfaces.

Pouring temperatures are checked by a disappearing filament optical pyrometer, and comparable figures are obtained from heat to heat. Stools sometimes are coated with tar, but as a rule they receive no preparation. The average stool life is 95 heats from plates 9 in. thick.

For any given teeming temperature, fast pouring rates produce ingots inferior in surface quality but with improved results in the mill.

Maintenance

Continuity of production depends largely upon the excellence of the mechanical, electrical and bricklaying maintenance organisations. Frequent inspection of plant is the backbone of good maintenance. Certain items of plant at Kemble are inspected daily. Once weekly all the furnaces, with auxiliary equipment, are carefully examined above and below floor level. An estimate of refractory and mechanical repair requirements is recorded and a copy forwarded to the foreman bricklayer and foreman mechanic, who then co-operate to have repairs effected without delay. This system works well, and assists the tonnage rate by eliminating unforeseen delays.

Accumulations in downtakes are removed at regular intervals, and slag-pocket deposits on fixed-fuel furnaces are removed at stipulated times.

(To be continued.)

The Institute of British Foundrymen

THE 40th annual general meeting of the above Institute has been arranged for June 26, 1943, at the Waldorf Hotel, Aldwych, London, W.C. 2. The business meeting will commence at 9-30 a.m., which will include the induction of officers, the President's address, and the awards of medals. Immediately following, Dr. S. F. Dorey will deliver the Edward Williams Lecture on "The Contribution of the Steel Founder to Marine Engineering."

The afternoon will be devoted to the presentation and discussion of technical papers, which are divided to form three sessions to proceed simultaneously. At Session A the following papers will be presented :—

"The Effect of Casting Conditions on the Properties of a Magnesium Die-casting Alloy," by Professor W. R. D. Jones, D.Sc.

"Report on a Copper-Antimony-Nickel Gear Alloy," by the Non-Ferrous Sub-Committee of the Technical Committee.

"A Method of Correlating Foundry Procedure and Quality of Castings," by H. G. Warrington.

Section B :—

"Some Aspects of the Production of Malleable Castings," by J. Roxburg

"The Spectrographic Analysis of Cast Iror," by F. B. Ling, J. McPheat, and J. Arnott, F.I.C.

"Meeting the Raw Materials Supply Position," a report by the Cast Iron Sub-Committee of the Technical Committee.

"Industry and Education," by G. L. Harbach and R. Horton.

Section C will be a joint meeting with the Iron and Steel Institute at which the following papers will be presented :—

"Hot Tears in Steel Castings," by Charles W. Briggs. This is the American Foundrymen's Association exchange paper.

"The Continuous Production of Manganese Steel Castings from the Tropenas Converter," by L. W. Bolton and J. Hill.

"A War-time Steel Foundry : Some Problems and Developments," by E. Derek Wells and A. Johnson.

Transformer Welding Equipment

THE General Electric Co., Ltd., has prepared a booklet describing their single and double operator portable welding sets. In addition to some useful notes on the subject of welding, this booklet describes the construction and electrical characteristics of the sets with useful information on their selection and operation. The subject is of vital importance to-day, and users of this equipment can obtain a copy of the booklet by applying to the G.E.C. through any of its branches.

Induction Furnaces for Melting Aluminium Alloys

For some time past developments in the production of aluminium alloys have been concerned with production technique rather than with the development of new alloys. Attention has been directed to the melting of aluminium and its alloys and an induction furnace, developed for this purpose, is described.

THE low-frequency induction furnace, largely used in the melting of copper and brass, has only recently been successfully adapted to the melting of aluminium and its alloys. In the development of aluminium alloy furnaces difficulties are encountered on account of the low density of the metal and the fact that the slags formed have no marked tendency to float, as in the case with heavy metals. Aluminium alloy slags tend to contaminate the melting channels of induction furnaces, and as such slags are non-conductors, and if any amount of them incrustate the melting channels, their resistivity increases and the power absorbed by the furnace tends to decrease and finally to stop. Such difficulties can be overcome by the proper design of the melting channels. The description, therefore, by M. Tama* of the development of an induction furnace for aluminium alloys is of outstanding interest, particularly at a time when aluminium and its alloys is being melted in large quantities for war purposes.

In a single-phase furnace, a primary coil, made of mica or glass tape insulated copper strip, is arranged around a transformer core, fabricated of high permeability silicon sheet steel. A transit tube surrounds the coil and insulates it from the lining of the furnace, and a blower and adequate cooling ducts are provided to keep the transformer winding and core cool during the furnace operation. When alternating current of standard frequency is passed through the primary coil a current is induced in the secondary loop or melting channel, and the amount of current so transmitted from the primary coil to the melting channel depends on the voltage and frequency of the current applied to the primary coil, on the number of turns on the coil, on the shape and size of the melting channel, and on the resistivity of the metal being melted.

The melting channel opens into the hearth containing the molten charge and the heat is generated only in the melting channel and transmitted rapidly to the hearth by virtue of electromagnetic forces which keep the metal in permanent circulation and creates ideal conditions of temperature gradients within the molten metal. The melting channel has two vertically straight legs which enter abruptly, without flaring openings, into the hearth, and this feature is important because the straight legs of the channel can easily be cleaned with straight bars during the melting operation without emptying the furnaces. The electromagnetic force is responsible for expelling small particles out of the channel into the hearth, and the slag is thereby brought to the surface, where it can be skimmed out from time to time. The lower horizontal portion of the melting channel can also be cleansed from time to time through two openings at the bottom, which are kept sealed by adequate plugs.

In operating the furnace the highest temperature is found at the bottom of the melting channels, slightly lower temperatures prevail in the hearth or crucible, and much lower temperatures are found in the atmosphere about the metal bath. If the molten charge in the hearth is kept at 705° C., the bottom of the channel attains a temperature of 750° C., while the air on the top of the bath shows only 510° C. The low coefficient of emissivity of aluminium alloys is responsible for these ideal conditions. In a fuel-fired or electric furnace, such as a reverberatory or electric resistance furnace in which the heat is transmitted by radiation, very high temperatures, over 1,100° C., must be kept above the melting chamber to ensure efficient heating. The low ambient temperature above the molten charge in the induction furnace helps in reducing the melting losses and the gas pick-up.

The absence of harmful gases is an important factor, as it has long been established that hydrogen and water vapour are the most harmful gases for aluminium alloys, and that the absorption rate increases with temperature. Hydrogen is not present in the atmosphere of an induction furnace, and although steam may be present in humid weather, the low temperatures prevailing above the melting chamber are not sufficient to dissociate the steam to form hydrogen. Gases may be brought into the bath with the scrap charged into the furnace, but they seem to be expelled to the surface by the permanent stirring of the melt. Repeated experiments have demonstrated beyond doubt that melts prepared in induction furnaces show a high degree of soundness.

The automatic stirring of the metal bath is a characteristic of all induction furnaces. In melting aluminium, the head of molten metal contained in the hearth must be high enough, so that the protective skin of aluminium oxide formed on the surface should not be broken. If this precaution is observed, the automatic stirring has only beneficial effects for the proper melting of aluminium alloys in so far as uniformity of temperature and of chemical composition are easily obtained within the charge. Another advantage of the electromagnetic stirring is that, in many cases, ingredients of high melting point, like copper or nickel, can be dissolved directly, thus eliminating the use of hardeners. Copper can be readily dissolved in the amount indicated by the constitutional diagram at a given temperature. This can be easily demonstrated by holding a copper bar in the aluminium bath when, due to the rapid movement of the aluminium, copper is dissolved before its melting temperature is reached.

In electric melting practice the production rate is usually expressed in terms of pounds per kilowatt-hour. The theoretical production rate for aluminium to a temperature of 730° C. should be equal to $1/0{\cdot}141 = 7{\cdot}1$ lb. per kw.-hour. The practical rate obtained in furnaces in operation up to the present is about 5 lb. per kw.-hour., which is equivalent to 70·5% efficiency.

1 *Metals and Alloys*, 1943, vol. 17, No. 2, pp. 326–329.

This efficiency can be improved by reducing the capacity of the furnace—that is, the weight of the charge to be tapped from the furnace—and at the same time increasing the power absorbed by the furnace. In the application of induction furnaces, therefore, careful consideration has to be given to the necessary weight of the charge. Furnaces of small tonnage provided with high-power ratings should be given preference whenever possible.

In the United States the trend of aluminium producers and of secondary melters is to deal with aluminium pigs of specified analysis ready to use in the furnace. The predominant practice is to melt those pigs in a reverberatory furnace and to ladle out metal in a molten state so as to bring a small charge to a holding furnace. By the use of an induction furnace, the holding furnace can be eliminated, the solid aluminium alloy pigs being charged continuously to small induction units and the molten metal discharged also continuously from these furnaces to the moulds. Adequate temperature controllers have been devised to keep the molten bath at the desired casting temperature within close limits, and automatic devices have been designed to charge the pigs continuously to the induction furnace, which by acting as a melting and a holding unit at the same time results in economy of operating costs and space.

New Spot Welder Model

AN interesting addition to the Holden and Hunt range of spot welders is their latest production, the fully automatic vertical sliding head air-operated spot welder, with standard transformer ratings of 25, 50 or 75 k.v.a. Other transformer ratings between 20 and 100 k.v.a. are available if required, and maximum electrode pressures of from 500 lb. to 2,500 lb., depending on the size of the machine. All models include air return to head—that is to say, pressure is applied to the weld and the electrodes opened by compressed air, so ensuring constant electrode pressure independent of any alteration in gap between the tips. Fig. 1 illustrates the 75 k.v.a. model, showing the exceptionally clean design and absence of all but essential external wiring and piping. In this illustration the switchgear cover is removed to show the ignition tubes, etc.

These spot welders are automatic in that once the operator presses the foot switch the control circuits take over and the machine is interlocked until the completion of the cycle of operations—i.e., squeeze, weld, forge and electrodes opened for the next operation. Initiation of the actual weld-time period is by means of a differential pressure switch which requires no adjustment for the various air pressures applied to the operating cylinder; it also ensures that correct pressure is applied to the electrode tips before heat is applied to the weld.

Air is supplied to the operating cylinder in sequence from the main shop air-line as follows:—Air purifier to remove condensate, H and H; adjustable air-line lubricator, H and H; pressure regulator, the control knob of which is placed immediately under the pressure indicating gauge, enabling adjustment of welding pressure between electrode tips to be speedily effected, and finally, solenoid operated air valve, the energising and de-energising of which is regulated by the electrical timing circuits.

The timing equipment supplied as normal standard is weld timer, B.T.H. Thyratron Contactor type, giving a range of from 5 to 500 cycles ($\frac{1}{10}$ to 10 secs.), forge and "time-off" being effected by adjustable magnetic-pneumatic relays with air diaphragms, the control knob of the former as in the case of the rheostat of the weld timer being mounted on the control panel. Setting of the "time-off" timer can be carried out by opening a hinged panel at the rear of the machine, this adjustment being but rarely required.

The control circuits described above are known as standard, but B.T.H. Ignitron contactors which give considerably increased advantages (especially in the case of models of 50 k.v.a. and upwards) over the normal

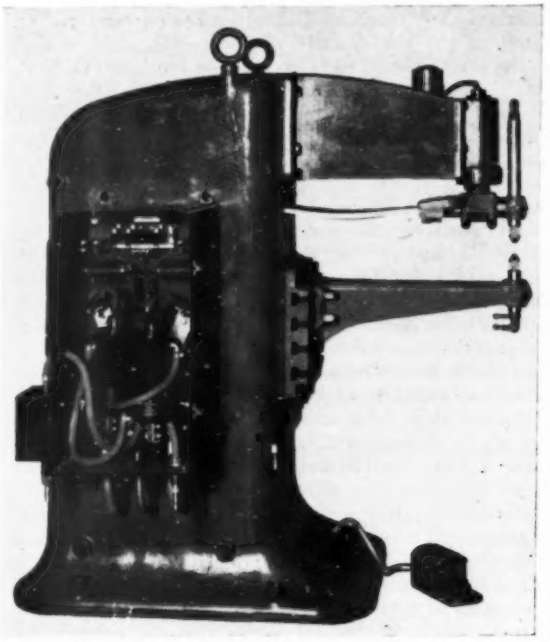


Fig. 1.—New 75 k.v.a. model spot welder with switchgear cover removed.

electro-mechanical contactors may be fitted as shown in the case of the 75 k.v.a. model illustrated.

Another alternative timing equipment fitted is the B.T.H. Woodpecker Ignitron control, giving repeated weld current impulses during the total heat period, which is a valuable feature where heavy gauge materials are to be welded. The larger models of this new range of H. and H. spot welders can also be supplied as light-duty press welders for projection welding, the specification being suitably modified and augmented to suit this duty.

Target for To-Morrow

A SERIES of books on post-war planning under the above title is soon to be published. Each book of the series will deal with a specific problem—health, nutrition, social security, and so on. The problem, and the main points of view about it, will be stated as objectively and scientifically as possible, with the minimum of partisan bias. Sir William Beveridge, Dr. Julian Huxley and Sir John Boyd Orr comprise the editorial board and Mr. Charles Madge is editor.

Cupola Melting of Cast-Iron Borings

With the present need for putting into useful service all available cast-iron scrap, the question of what to do with the large quantities of borings has aroused considerable attention. A United States process is described and comparison is made with a similar process which has been in operation in a British foundry during the last four years.

IN this country and in the United States the problem of disposing of cast-iron borings and turnings, thousands of tons of which are produced in engineering and machine-shops every year, is one which has aroused considerable attention during recent years. Until the present shortage of almost all types of scrap developed, each producer had his own particular means for the disposal of such borings. The shortage of scrap, however, has redoubled the efforts of iron foundries having machine-shops to use the borings produced from their own castings, and a number of methods have been suggested in an attempt to convert such scrap to some usable condition. Briquetting produces a product suitable for charging into the cupola and large producers have completely equipped salvage plants for the briquetting of both iron borings and steel turnings. The average producer, however, does not accumulate sufficient tonnage to justify the expense of installing even a small briquetting plant, and the description of a method for the direct cupola melting of loose iron borings, whereby the average iron foundry may melt all the iron borings produced in its plant, by S. L. Feduska,* is therefore of interest to all such producers.

The development of this method for the melting of loose cast-iron borings by charging directly into the cupola arose from an urgent need for a certain class of chromium-molybdenum iron scrap, low in phosphorous, for the manufacture of cast-iron rolls. Worn-out rolls not being available for scrap, experiments were made with borings and turnings, whose density varied from 35 lb. to 95 lb. per cub. ft., by charging them directly into a 45-in. cupola. The actual experiments consisted in varying the size of the charge, the blast pressure, and the height of the stock. The amount of turnings melted varied from 2,000 lb. to 14,000 lb. per day. The first two tests, consisting of single melts of 4,000 lb. and 2,000 lb., were made at the end of a daily cupola run to determine whether the borings could be melted without going " up the stack." In these first tests a low blast pressure was used, but in succeeding tests the pressure was gradually increased to that used in melting a normal iron charge.

These first tests were followed by other tests, the purpose of which was to continuously melt larger amounts of iron borings. In the first attempt to melt continuously 14,000 lb. of borings, some difficulty was experienced in melting, and observation showed that a normal blast had difficulty in penetrating the stock charge, requiring a tremendous increase in blast pressure. It was therefore found necessary to decrease the charge size and also to decrease the stock height. The effect of these changes resulted in an improvement of melting rate, and rendered it possible to melt continuously at normal blast pressures. In succeeding tests the melting rate was gradually improved, and continuous melting rates equal to 8 tons per hour in a 45-in. cupola were obtained.

The best practice was found to be melting with a blast pressure of 6 oz. to 10 oz., maintaining a stock height of 4½ ft. to 5½ ft., and having charges consisting of 15,000 lb. to 2,000 lb. of turnings; 150 lb. to 200 lb. of coke and 35 lb. of limestone. The amount of metal lost was less than 5%, this value being determined by finding the difference between the weight charged, turnings or borings, and the resulting pigged iron. The metalloid losses on melting the turnings of an iron having an approximate composition of 3·14% total carbon, 1·09% silicon, 0·44% manganese, 0·82% chromium, 0·35% molybdenum was 25% chromium, 25% silicon, 50% manganese, and 25% molybdenum. In one case only the carbon content dropped to 2·56%. In the best continuously melted heats of about 5 tons, the carbon ran higher than those of the charged turnings. When the turnings had a carbon content of about 3·14%, the carbon contents of continuously melted turning heats ran 3·30 to 3·60%, indicating that some carbon had been absorbed from the coke.

The volume of slag in the tests carried out under the best conditions appeared to be more than that obtained from cupola melts of the same size, and on examination was found to have a very low density when compared with the slag obtained from regular iron melts. The slag was of such a fluid nature that it was aerated to a foamy mass on leaving the slag hole. Examination of the cupola lining after each test showed that slightly more than the usual erosion of the lining occurred in the early tests, but that later tests did not show any more lining erosion than that expected in melting regular iron heats of the same size. In any case, the amount of lining erosion, even under the worst melting conditions was not considered excessive.

The process described was developed in a foundry in the United States, and is compared with a similar process for the direct cupola melting of cast-iron boring practised for the past four years in a British foundry. In the latter process, 6 to 7 tons of loose iron borings were continuously melted each day in a 33-in. balanced blast cupola. The coke bed was kindled and built up in the usual way and two charges consisting of 500 lb. of cast-iron scrap were placed on top of the coke and the blast turned on. The level of the stock in the furnace was maintained by charges made up of 400 lb. loose iron borings, 200 lb. pig iron or scrap, 100 lb. of coke, 75 lb. of limestone, and 50% of silica sand. The height of the stock inside the furnace was found to be of great importance because, if the stock line was too high, the blast was unable to penetrate and the cupola eventually stopped melting. A stock height of 5 ft. or lower gave the best results, and the charging was such to maintain this height. The highest pressure, about 20 in. of water, of which the fan was capable was used. The melting rate was about one-third normal, and an output of 1¼ tons per hour was obtained from the 35-in. diameter cupola.

The limestone addition in the British practice was

Metals and Alloys, 1943, vol. 17, No. 2, pp. 316–319.

considerably heavier than that used with normal cast-iron charges, and the silica sand was added to neutralise the large lime addition. The larger amount of lime was added with the object of obtaining a large volume of slag to dilute the iron oxide produced from the borings. The weight of the slag produced was about 22% of the weight of the metal melted, and the total iron loss was found to be less than 3%. The lining erosion, although more than would be obtained in normal cast-iron practice, could not be considered excessive. The typical analysis of the iron produced, which was cast into pigs, was 2·3% total carbon, 0·80% silicon, 0·25% manganese, and 0·15% sulphur.

Using a 45-in. diameter cupola instead of a 33-in. diameter cupola, a melting rate of 6 tons of borings per hour was obtained with a charge consisting of 800 lb. of borings and 400 lb. of pig iron or scrap.

Besides making synthetic pig iron or foundry scrap, cupola melting of loose cast-iron borings and steel turnings may be used in a process for making synthetic steel scrap by melting the loose turnings in the cupola, desulphurising in the fore hearth, running into a mixer, reducing the carbon and metalloids in a Bessemer converter, and dephosphorising in the ladle. Using this procedure, it is believed that a large producer of cast-iron borings and steel turnings might make synthetic steel scrap at a cost lower than that of steel scrap purchased on the market.

The Midland Laboratory Guild

On January 22, 1918, at the Queen's Hotel, Birmingham, an informal meeting was held of the representatives of seven well-known companies in the local non-ferrous metal trade, together with the late Mr. W. R. Barclay, who was then chief metallurgist to the Ministry of Munitions. The object of the meeting was the foundation of a co-operative laboratory, under the guidance of the Ministry, to serve the whole group of companies present, and any others who might join them from time to time. The expense and difficulty of starting a separate laboratory in the works of each company had proved to be insuperable, and Mr. Barclay conceived the idea of a co-operative laboratory.

On May 24, 1918, the Midland Laboratory Guild Limited was incorporated, and in June the late Mr. G. H. Dugard was elected chairman, an office he continued to hold till the reconstruction of the Company in 1928. The laboratory was started in premises in King Alfred's Place, Birmingham, under its first scientist-in-charge, the late Mr. Seymour Pile, and with an advisory panel which included such distinguished men as the late Dr. C. T. Heycock, of Cambridge, Mr. W. R. Barclay himself, and Dr. C. H. Desch, F.R.S.

Mr. Pile, whose originality and resource contributed in a great measure to the initial success of the Guild, and who steered it through many early difficulties, set up a high standard of service and accuracy and laid down lines along which, to a great extent, it has since developed. In June, 1922, he retired from the service of the Guild, being succeeded by the present scientist-in-charge, Mr. R. G. Johnston.

Unforeseen difficulties, arising from restrictive Articles of Association which had been dictated by the Ministry of Munitions, without apparent thought of post-war conditions, led in July, 1928, to a reconstruction of the Guild as the Midland Laboratory Guild (1928) Limited, with a membership which had increased since its inception. In August of that year Mr. G. H. Dugard resigned the chairmanship, though till his death in 1932 he remained on the Board. Mr. W. H. Henman was elected to the chair, which position he has held with conspicuous success to this day.

The experiment of co-operation in running a central laboratory by companies otherwise unassociated, and even competitive, has proved a great success. It has made possible a funding of experience and a breadth of service which would be beyond the reach of any one company, unless it were of great size. The service of the laboratory has, in addition, been made available to non-members of the Guild, who are served on the footing of clients, and it is being utilised by many such. Since the beginning of the present war the activities of the Guild, both for its members and for its clients, have undergone great expansion, and the entire capacity of the service is taken up in work necessary in the execution of contracts for Service requirements.

The Guild specialises in the testing and control of non-ferrous metals and alloys, and in its consultative service. It is well equipped also for physical testing and photomicrographic work. It is thought that although co-operative laboratories are perhaps still few, the outstanding and consistent success of the Midland Laboratory Guild over the last 25 years may stimulate interest for the carrying on of similar lines of undertakings, even if dissimilar in character.

Grain Size in Tin Alloys

THE effect of grain size on the physical properties of metals has been studied by various research workers. Interesting light on the part played by grain size in the properties of pure tin and tin-rich alloys is given in the Tin Research Institute's publication No. 118, which is a reprint of a paper by Dr. W. T. Pell-Walpole in the *Journal of the Institute of Metals*.

Dr. Pell-Walpole's investigations show that variations in tensile strength are controlled by the number of grains in the cross-section, and not by the absolute grain size. In rolled material, an increase in tensile strength is observed as the number of grains in the cross-section increases from 1 to 20 or 30. Elongations are constant over this range. A further reduction in grain size produces only a slight increase in tensile strength, but elongations increase rapidly.

The discontinuity in the relationship between tensile strength and grain size is shown to be caused by the corresponding discontinuity in the extension/grain size curve, the latter having some fundamental cause. The calculated true stress was found to vary continuously with grain size.

With chill-cast specimens, consisting entirely of columnar grains, tensile strengths increase continuously with reduction of grain size, and elongations are either constant or increase only slightly. Copies of this publication may be obtained free of charge from the Tin Research Institute, Fraser Road, Greenford, Middlesex.